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TRANSACTIONS

American Society for Steel Treating

VOL. XV

APRIL, 1929

NO. 4

SURFACE HARDENING OF SPECIAL STEELS WITH AMMONIA GAS UNDER PRESSURE

BY RAYMOND H. HOBROCK

Abstract

Samples of special steels have been nitrided with ammonia gas under pressure and the nature of the case so produced has been examined. The results show that an increase in the pressure of the ammonia gas results in a decrease in the surface hardness but greatly increases the depth of case produced in a given time. Curves are presented correlating hardness, time of treatment and depth of the case. In a theoretical discussion of the results the author proposes an explanation of the phenomena observed.

NITRATION

WITHIN the last few years there have been introduced special alloy steels that are particularly suited for surface hardening by the use of ammonia gas at rather low temperatures. Dr. Adolph Fry of the Krupp Aktiengesellschaft, Essen, Germany, has experimented extensively with the compounds of nitrogen and iron, has largely developed the temperature-composition diagram for these compounds and has been largely responsible for the successful commercial production of the alloys that are finding increasing use in the European and American industries.

The physical properties of these alloys in the untreated condition have been determined by Defries of the Ludlum Steel Company and by Homerberg and Zavarine of the Massachusetts Institute of

A paper presented before the tenth annual convention of the society held in Philadelphia October 8 to 12, 1928. The author, Raymond H. Hobrock, was associated with the Engineering Experiment Station, Purdue University, Lafayette, Indiana. Manuscript received June 8, 1928.

Technology and others as listed in the attached bibliography. The properties of the case have also been determined by various workers. However, the nitration process has generally been carried out at atmospheric pressures and no previous attempts have been made to examine into the nature of the case that might be produced by the nitration of the special steels under pressure.

It is the object of this paper to set forth the results of an examination made in the Department of Practical Mechanics of Purdue University, into the relations existing between pressure of the gas (NH_3), hardness, penetration, and hardness gradient for one typical steel.

EXPERIMENTAL

In this investigation the three fundamental variables in the treatment of the samples might be (1) the temperature, (2) the time of treatment, (3) the pressure of the ammonia gas. However, the temperature at which the treatment is best effected has been studied in previous researches, and for this experiment, has been maintained constant at 875 degrees Fahr. The time of treatment has been allowed to vary from two hours to 100 hours and the pressure of the ammonia gas has varied from atmospheric pressure to 600 millimeters of mercury above atmospheric pressure.

The alloy used in these experiments had a composition as indicated in the following limits:

	Per Cent		Per Cent
Carbon	0.38—0.43	Aluminum	1.00—1.25
Silicon	0.20—0.30	Molybdenum	0.15—0.25
Manganese	0.40—0.60	Nickel	0.30—0.60
Chromium	1.60—1.80		

The specimens were cylindrical and were cut from a rod of the material 11/16 of an inch in diameter. The two ends were ground smooth and parallel to each other on a surface grinder so that when finished the cylinders had a height of 0.5 inches. Previous to the finishing operations the samples were annealed for ten hours at a temperature of 1000 degrees Fahr. in order to remove rolling and machining strains which might later cause distortion.

The apparatus is indicated diagrammatically in Fig. 1. It consists essentially of a tank of liquid ammonia, a needle valve for controlling the velocity of the ammonia gas escaping from the cylinder, a fused quartz tube contained in the electric furnace, a

thermocouple for the regulation of the furnace temperature, a mercury manometer on which the pressures above atmospheric may be read, and a mercury column for maintaining the pressure inside the tube constant. The other apparatus shown in the diagram serves in preventing the escape of ammonia into the atmosphere and in observing the rate of flow of the gases through the apparatus.

Nitration of the Samples

After the temperature and the pressure of the ammonia had been adjusted the samples were introduced into the furnace and

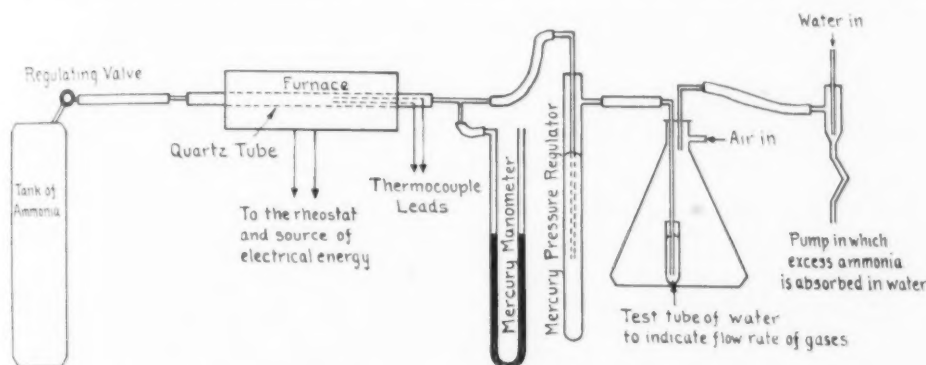


Fig. 1—Schematic Diagram of the Apparatus for Pressure Nitriding.

treated continuously for the various lengths of time. They were removed after the expiration of the required time and were allowed to cool in a desiccator containing concentrated ammonium hydroxide in the bottom.

Testing of the Samples

The surface hardness of the specimens was then determined with a Shore scleroscope. Many readings made near the center of both flat faces of the samples assisted in determining an average result and in ascertaining which face gave the most consistent readings. The samples then had 0.001 inch ground from the face selected on a surface grinder which was so arranged that the new surface produced was always parallel to the opposite face. The hardness of this new surface was then determined with the scleroscope and the exact depth of the first cut determined by measurement with micrometer calipers. Subsequent procedure was similar except that the depth of the cuts was increased to about 0.002 inch. This

grinding and hardness testing was continued until the hardness numbers were the same as the hardness numbers of the material before the treatment.

An experiment was also made to determine whether or not the small surface scale produced in the nitration process effected the hardness numbers as determined with the scleroscope. Several of the specimens were tested as they came from the furnace, polished with Tripoli powder to remove the scale and were then tested again. No difference in the hardness numbers was observed and accordingly the remainder of the specimens were not polished before being tested for hardness.

THE CURVES

The curves numbered from Fig. 2 to Fig. 5 show the change in hardness with the depth, i. e. they are hardness gradient curves.

An examination of these curves will show that when the pressure of the gas is increased the time required for the establishment of a given case depth is reduced. Thus when the specimens were treated at a pressure of 5 millimeters for 100 hours there resulted a case of about 0.027 inch in depth. If the pressure be increased to 602 millimeters the steel will require treatment for about 50 hours in order to establish this same case depth (see curves Figs. 2 and 5), thus decreasing the time of treatment 50 per cent. The use of the higher pressure, however, results in a somewhat lower surface hardness in the steel under examination (this may be corrected—see theoretical discussion). The treatment of these steels under pressure, therefore, results in great saving in time and considerable saving in the cost of the treatment.

Curve (Fig. 2) indicates what might be expected when this kind of steel is treated with a pressure of ammonia just slightly above atmospheric. The 5-millimeter pressure indicated was due to the small column of water through which the gas bubbled at the outlet end of the apparatus in order to indicate the rate of flow of the gas.

Curve (Fig. 6) shows the way in which the surface hardness varies with the time of treatment for any particular pressure, while Fig. 8 shows the relation between the depth of the case produced and the time of treatment for the minimum and for the maximum pressures dealt with in these experiments. The depth of the case

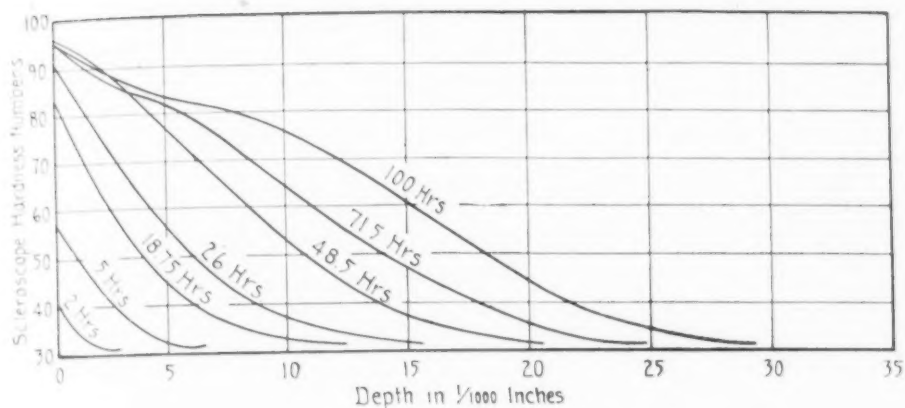


Fig. 2—Hardness Depth Curves for Nitrided Specimens. Temperature 875 degrees Fahr. Ammonia Pressure 5 Millimeters above Atmospheric Pressure.

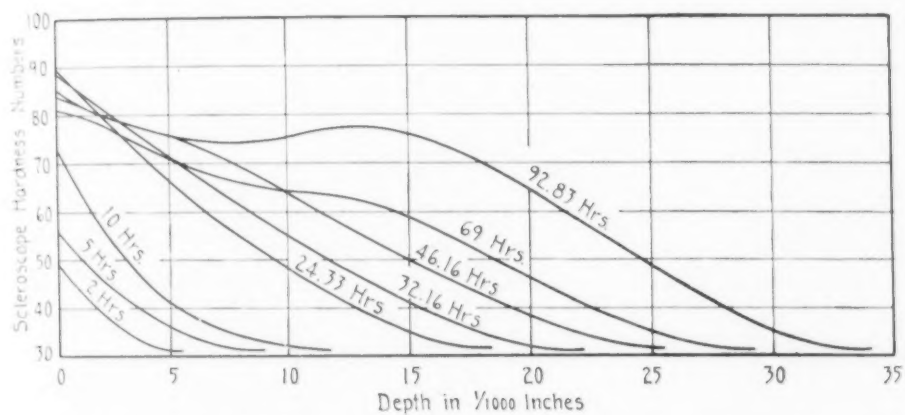


Fig. 3—Hardness Depth Curves for Nitrided Specimens. Temperature 875 degrees Fahr. Ammonia Pressure 202 Millimeters above Atmospheric Pressure.

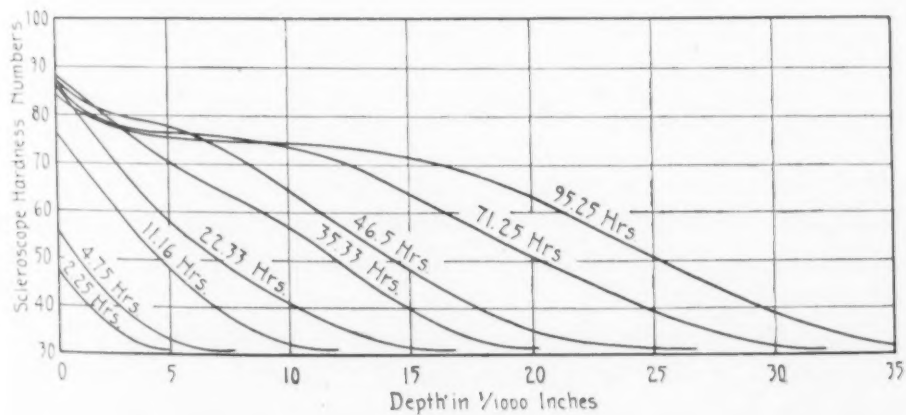


Fig. 4—Hardness Depth Curves for Nitrided Specimens. Temperature 875 degrees Fahr. Ammonia Pressure 402 Millimeters above Atmospheric Pressure.

was obtained from the gradient curves and is defined as the depth at which the material reached a hardness of 32. The hardness of the original material was 31 but the higher value of 32 was selected

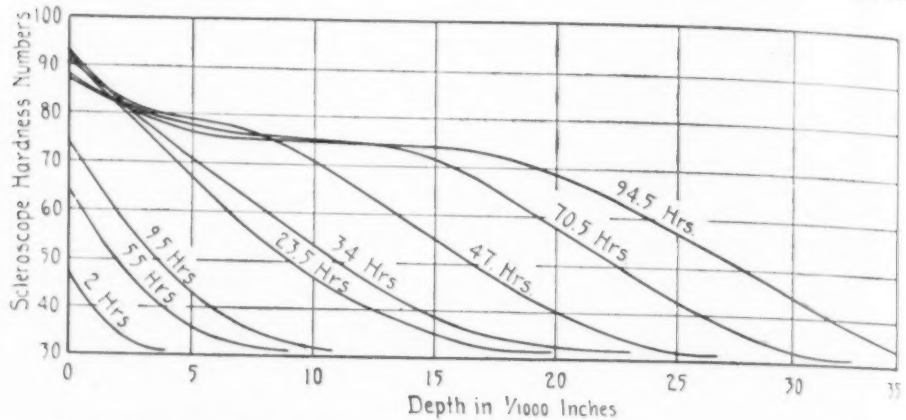


Fig. 5—Hardness Depth Curves for Nitrided Specimens. Temperature 875 degrees Fahr. Ammonia Pressure 602 Millimeters above Atmospheric Pressure.

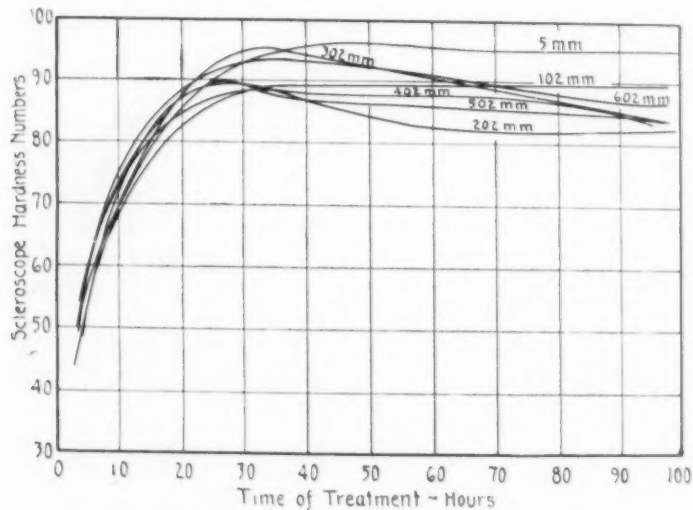


Fig. 6—Relation Between Surface Hardness and Time. (Pressure Constant).

in this case because it is defined much more readily from the curves. The depth value for all of the pressures intermediate between 5 millimeters and 602 millimeters fall between the two curves in quite a regular fashion.

The curves numbered from Fig. 9 to 12 show how the hardness varies with the time of treatment at the various depths when the pressure is constant and of the magnitude indicated on each curve. Samples treated at pressure of 100, 300 and 500 millimeters were

also examined and exhibited the same tendencies that are shown in the curves presented.

RESULTS

A general consideration of the results seems to indicate that the increase in pressure,

1. Does not greatly change the time required for reaching the maximum surface hardness
2. Does tend to decrease the hardness at the surface if the treatment is continued for longer than about 25 hours
3. Does considerably increase the depth of the case
4. Does decrease the average hardness gradient
5. Does permit the establishment of a high hardness at almost any depth depending on the pressure of the gas and the time of the treatment.

THEORETICAL CONSIDERATION OF RESULTS

The great hardness of the cases produced by the nitration of the special steels cannot be explained alone by a consideration of the iron nitrides, even though the formation of such nitrides increases the hardness to a marked degree. If the phenomena be examined in the light of the hardness theory of Jeffries and Archer we might explain readily not only the extraordinary hardness produced but also the conditions which we have observed in the specimens treated under pressure for various lengths of time.

Thus, the change in the hardness produced by the nitration of electrolytic iron might be due to a change in the absolute cohesion

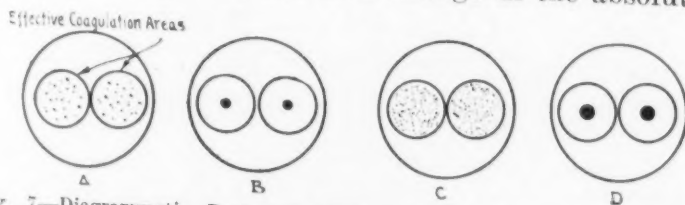


Fig. 7—Diagrammatic Drawings Representing the Formation of Key Particles of Different Sizes of Coagulation.

of the iron itself or it might be due to the formation of interference particles which prevent movement of the grains along the slip planes. From the nature of the crystals of the nitrides of iron which have been observed under the microscope it seems that the hardness might be best accounted for by considering it largely as

an interference phenomena. The heterogeneous and interlocking nature of these iron nitride crystals (which are visible under the microscope) are in some respects comparable to interlocking dendrites and act as an effective interference medium, not because of their dispersion in the slip planes themselves, but because of this

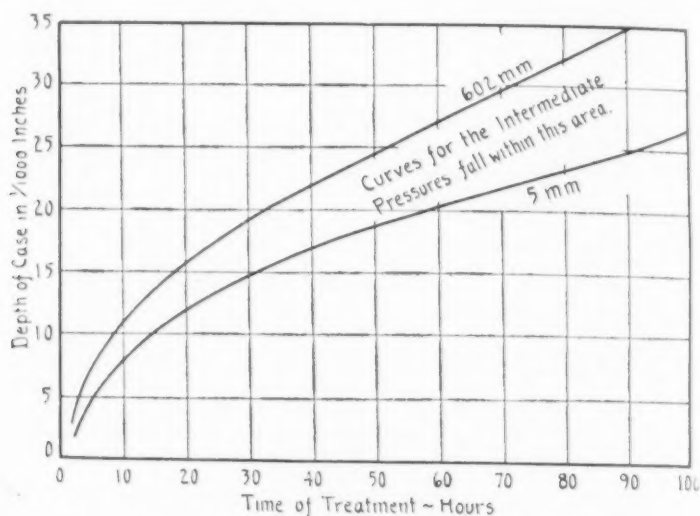


Fig. 8—Relation Between Case Depth and Time.

network formation. Such needlelike crystals of iron nitride frequently extend through other grains and in this way assist in the formation of smaller grains which in itself might account for some of the increase in hardness. The order of magnitude of this increase in hardness of electrolytic iron may be seen from Table I in which is also listed a hyper and a hypoeutectoid steel for comparison.

The carbon steels did not show the great increase in hardness that the iron did, as might be expected from the fact that the steels already contained a material in the form of cementite which ef-

Table I
Hardness Values

Material	Carbon Per cent	Brinell Numbers		Increase
		Before nitriding	After nitriding	
Electrolytic Iron	0.05	90	140	50
Carbon Steel	0.62	215	234	19
Carbon Steel	1.27	278	285	7

effectively interfered with deformation by the formation of network structures (pearlite, free cementite) in the grain boundaries. It seems from this that no great portion of the hardness of these special steels is due to the formation of the iron nitrides especially

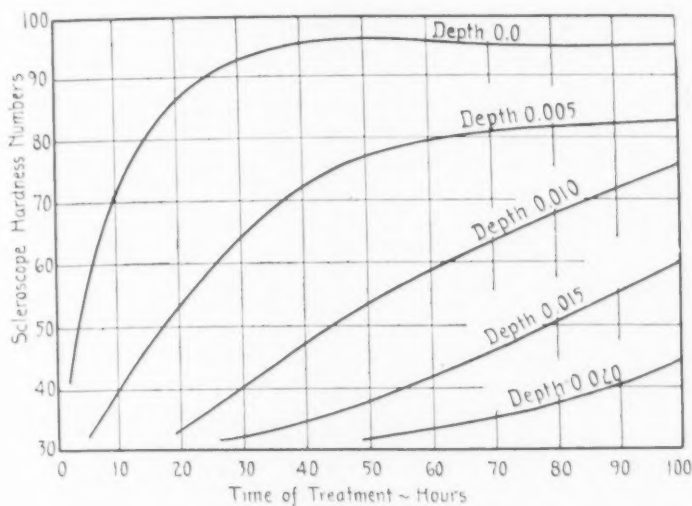


Fig. 9—Relation Between Hardness and Time. Pressure 5 Millimeters above Atmospheric Pressure.

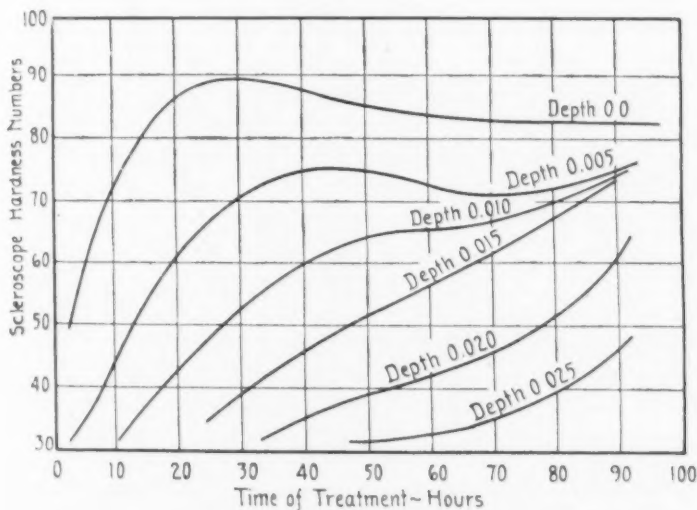


Fig. 10—Relation Between Hardness and Time. Pressure 202 Millimeters above Atmospheric Pressure.

when the special steels themselves have carbon contents of the nature indicated previously in the analysis of the steel investigated. It is necessary therefore, to seek an explanation in the effect of the nitration on the alloying elements.

One of the alloying elements contained in almost all of the steels intended for nitration is aluminum. Other elements are sometimes added in order to improve the qualities of the metal that

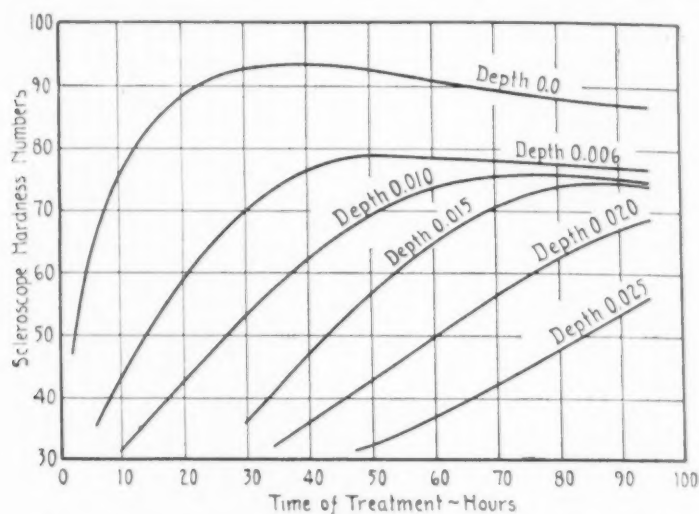


Fig. 11—Relation Between Hardness and Time. Pressure 402 Millimeters above Atmospheric Pressure.

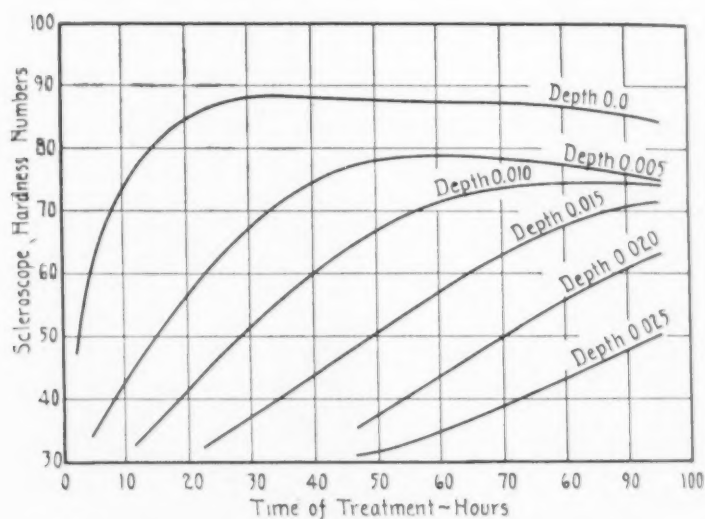


Fig. 12—Relation Between Hardness and Time. Pressure 602 Millimeters above Atmospheric Pressure.

will constitute the core, or in order to effect in some way the case formed.

An aluminum nitride Al_2N_2 is well known. It is a crystalline material with a melting point of 3900 degrees Fahr. (2150 degrees Cent.) [compare platinum, melting point 3190 degrees Fahr. (1755

Table II

Nitrides Made	Material Used	Treatment With	Time (Hours)	Temp. (° C.)	% of N After Treatment
Iron Nitride	Electrolytic Iron	Ammonia	72	600-650	6.7
Manganese Nitride	Manganese (87.6% Mn)	Ammonia	48	600	11.8
Chromium Nitride	Ferro-chrome .03% C 61.3 % Cr	Ammonia	48	650	8.2
Chromium Nitride	Pure Cr	Ammonia	48	650	10.6
Tungsten Nitride	Ferro-tungsten 84% W	Ammonia	48	650	3.6
Titanium Nitride	Ferro-titanium 23% Ti	Ammonia	48	650	4.1
Vanadium Nitride	Ferro-vanadium 49.8% V	Ammonia	62	650	7.9
Aluminum Nitride	70% Al powder 30% Al ₂ O ₃	Ammonia	½	1150	14.6

degrees Cent.)] Some of this material may be formed in the nitration process, however no experiments have confirmed the existence of all the nitrides of aluminum in this form.

In order to examine the properties of alloying materials that might be of use in the steels intended for nitration, Dr. Fry sub-

Table III

Nitrides Tested	% N before Treatment	Heat Treatment at Temperatures (° Cent.)							Total N given off	N in Residue
		20°	400°	500°	600°	700°	800°	900°		
Iron	6.7	tr	tr	6.64	2.37	tr	0	0	6.88	.005
Mn	11.8	tr	0.108	1.47	5.47	0.52	1.12	1.10	9.804	1.9
Cr (Fe)	8.2	0.06	0.289	0.816	0.165	0.358	1.88	1.265*	4.8	4.1
Cr	10.6	0	0	0	0	tr	0.605	3.06	4.1	6.8
W (Fe)	3.6	0	tr	0.31	0.34	1.40	0.75	0.06	2.9	0.6
Ti (Fe)	4.1	tr	0.656	0.110	0.104	0.045	0.022	0.9	3.4	
V (Fe)	7.9	0.12	0.227	0.847	0.16	0.185	0.094	0.021	1.65	6.3
Al	14.6	0	0	0	0	0	0	0	0	14.6

Nitrogen in the residue determined by analysis. The difference in the percentages of nitrogen in the untreated samples and the sum of the nitrogen given off plus the nitrogen left in the residue is probably due to the lack of homogeneity of the samples under test.

*This sample was heated to 1100 degrees Cent. The total nitrogen given off and the nitrogen in the residue are determined accordingly.

jected a number of alloys and mixtures to the nitration treatment and then examined the properties of the materials in several ways. The above table from his works shows the results of his experiments. (See Table II.) From the table it is apparent that the nitrides of aluminum are formed when treated with ammonia gas at a temperature of 2100 degrees Fahr. (1150 degrees Cent.), for only one-half hour.

Table III, shows the results of heating the nitrides so formed to various temperatures. The aluminum nitride showed no measurable decomposition at even the highest temperature. Thus two of the reasons for wide use of aluminum in these steels becomes apparent, viz; they can be formed by treatment with ammonia gas, and they are very stable.

In order to explain the effect of the aluminum on the hardening of the steels when nitrated it is only necessary to imagine the formation of aluminum nitrides from the aluminum that exists in solid solution in the steel. If the nitrogen is brought into the steel and is in fact absorbed by the steel, and so comes into intimate contact with the aluminum, the formation of the stable nitrides of aluminum is easily explained. It may in fact be compared to the reaction between a water soluble chloride and silver nitrate to form the very insoluble silver chloride. The formation of such aluminum nitrides from the solid solution does not in itself explain the increase in hardness because the particles so formed would be so minute that they would not effectively act as interference or key particles. But if these very minute particles can be caused to increase in size either by the formation of more nitride or (what is more likely) by the segregation of the nitride particles, then they may act effectively in the interference of slip.¹ However, to effectively interfere with the slip, the particles must be neither too large nor too small. If the particles are too small they will fit in between the slip planes and offer comparatively little interference. Thus particles smaller than the dimensions of a unit crystal of the surrounding iron might fit into the structure and not effect the properties of the material in any way except in its effect on the absolute cohesion. If, on the other hand these key particles have dimensions somewhat greater than the unit crystal, then forces acting along the planes of slip will be resolved into several forces acting at angles other than parallel to the slip plane, and so the force acting to cause shear along the plane will be changed in direction. But if the size of the particles approach the size of the grains themselves, then they will offer a minimum interference to the slip since the formation of particles of such size would represent

¹The change in hardness of certain types of duralumin with time and at atmospheric temperatures is an example of the coagulation phenomena and is, in fact, the phenomena for which the theory of Jeffries and Archer was first formulated. Quenched samples of duralumin showed an increase in hardness upon standing. This hardness reached a maximum after eight days and then decreased to nearly its original value. See *Chemical and Metallurgical Engineering*, Vol. 24, No. 24, p. 1057, for a discussion of this theory.

the disappearance of many small particles, and since occasional particles of such great size (even though no slippage occur within the particle) would obviously have little effect on the slippage of the whole of the material.

If the results of these experiments be considered in the light of this theory the various phenomena may be readily explained. In the case of the nitration of the steel at atmospheric pressure, there results, after about 45 hours of treatment, a surface hardness which changes very little with greater lengths of treatment. However, when the treatment is continued for 100 hours there is a measurable, though slight decrease in this surface hardness. This may be seen from curve, Fig. 9. With increasing pressure of the gas this decrease in the surface hardness is much more apparent and of considerably greater magnitude. (Figs. 9 to 12) This behavior might be explained by assuming that with increasing gas pressure the penetration of the nitrogen would be greater and the formation of more particles would result. The subsequent coagulation of these particles into effective interference particles, results in the formation of particles of such size as to offer an interference which is not the maximum. In other words the interfering particles have become a little too big for optimum interference. In the treatment at atmospheric pressure for 45 hours there seems to have been formed just sufficient particles of the right size to cause this maximum interference. The formation of more particles does go on with increasing time of treatment however, even under atmospheric pressure and results in the slight decrease in the hardness noted after the 100-hour treatment. Increasing the pressure of the gas then results in accelerating the formation of particles by increasing the concentration of the nitrogen to react with the aluminum.

If the amount of aluminum in the solid solution be increased then the amount of nitride particles formed would be increased, and, when the coagulation process had reached an equilibrium at a certain temperature, the key particles produced by this coagulation would be of greater size and there would result an actual decrease in the hardness of the material. Indeed this has been observed in experiments with an alloy steel which contained 3.28 per cent aluminum and one which contained 2.50 per cent aluminum. After both of these steels had been nitrated under the same conditions, the one which contained the 3.28 per cent of alu-

minum showed an increase in Brinell hardness of 215 while the other which contained but 2.50 per cent aluminum showed an increase of 340. See Fig. 7. These figures represent the formation of key particles of different sizes by coagulation. The small circles represent the area in which coagulation of the small particles to a larger particle can occur (constant at constant viscosity, temperature, etc.) "A" shows two of such areas with a small amount of the highly dispersed aluminum nitride within the areas. "B" shows these same areas after coagulation is complete. "C" and "D" show this same coagulation when there are more of the dispersed particles of the nitride available within the coagulation area. The particles resulting in "D" are larger than those in "B." In order to arrive at the maximum hardness there should be just sufficient dispersed particles within the area of coagulation to form key particles slightly larger than the unit crystals of the iron.

If the amount of aluminum in the alloy is constant and if the coagulation of the nitrides is allowed to take place at the same temperature, and if all of the aluminum is changed to the nitride in any given volume of the material, then the size of the coagulated particles should approach some constant size—which means that the nitrified steel itself should approach some constant value in hardness. That this is true is evident from the hardness gradient curves (Figs. 2 to 6)—especially those in which the samples were treated under pressure of the gas. All of these curves (which were made on the same kind of steel and therefore should have a uniform aluminum content) show a decided tendency to flatten out in the neighborhood of a scleroscope hardness of about 73. In some of the specimens this hardness was maintained for considerable depths. This hardness might then represent the hardness due to the size of the key particles formed from a definite amount of aluminum nitride which has been permitted to coagulate at a definite temperature (875 degrees Fahr. in this case) until equilibrium is reached. The increase above this value in the hardness at the surface might readily be explained by the formation of the iron and other nitrides in this region. The curves showing the relation between the hardness numbers at various depths and the time of treatment also show this convergence to a hardness value near to 73.

From the results of these experiments, as well as the work of others, (which seem to largely confirm the explanation of the hard-

ness of these alloys which has been offered) we might deduce the following:

1. An alloy intended for the establishment of a hard surface of no great depth might contain much more aluminum than an alloy intended for use where a deep case is required.
2. Where high surface hardness and deep penetration are to be obtained by nitration under pressure the amount of aluminum in solid solution in the alloy must be accurately controlled.
3. The increase in hardness due to the nitration of the alloy steels is due in part to the formation of a network of crystals of iron nitrides and in part to the formation of slip interference particles of aluminum (and other) nitrides.

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DISCUSSION

Written Discussion: By R. E. Bissell, Thompson Products, Inc., Cleveland.

I consider that Mr. Hobrock has performed some very interesting experiments relative to the relation between the hardness and the depth of penetration as developed by nitrogenizing of "Nitr alloy" steel. The charts he presented are of great help to those who are nitrogenizing commercially.

There may be some criticism as to the application of the scleroscope in determining hardness. It is, however, at least temporarily a very useful instrument for this purpose.

I believe Mr. Hobrock should go further in determining some practical method and apparatus whereby the benefits of nitrogenizing under considerable pressure can be taken advantage of.

Written Discussion: By J. D. Hoffman, Purdue University, Lafayette, Indiana.

One of the most important elements at the disposal of the production manager is *time*. If, as seems quite probable, the use of nitrated steel in special machinery might increase quite rapidly, any process that could shorten the time of nitration would be a distinct advantage to all parties concerned.

The project as set forth in this paper was first conceived to see if the process of nitration could be speeded up, and second, to make a study of the effect of such nitration on the physical properties of the metal.

Although not sufficiently versed in the science of chemistry and metallurgy to comment on this part of the work, I am pleased to mention the painstaking care exercised throughout the investigation in making plans for and executing the work so as to give, as near as possible, reliable results.

Written Discussion: By G. M. Eaton, Molybdenum Corp. of America, Pittsburgh.

The object of the paper is to set forth the results of an examination into the relation existing between pressure of gas, hardness, penetration and hardness gradient for one typical steel. The hardness was tested by means of the scleroscope. We might question the use of the scleroscope in research involving hardness determination on a steel of uniform hardness. We will, however, waive this point and focus on the fundamental reasons why the scleroscope results demand careful analysis when dealing with the hardness characteristics of a material whose hardness varies in a direction normal to its surface.

Fig. 1 shows a family of hypothetical hardness penetration curves. Curve A shows a steel of uniform hardness throughout, while curves

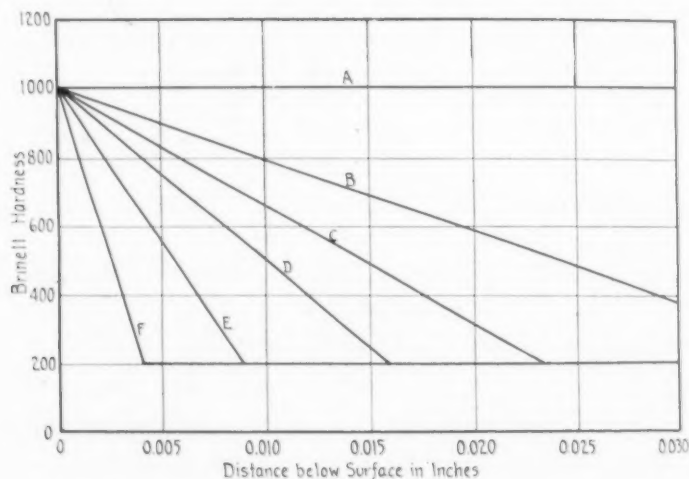


Fig. 1—Hypothetical Hardness-Penetration Curves.

B, C, D, E and F show hardness gradients dropping down with increasing rapidity to a uniform core hardness. The surface hardness of all six is identical. Anyone who has investigated the depth of the effect of the scleroscope impact will see at a glance that the scleroscope readings will be radically different on the outer skin of all six samples. The material represented by curve A will have the highest hardness reading and the readings will fall off successively to a lowest reading on curve F. In our study of the hardness penetration characteristics of nitrided cases, we have found no absolutely ideal way for measuring hardness. We consider

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the Herbert pendulum with the sapphire or diamond ball capable of the highest degree of technical accuracy and we consider the Vickers hardness tester as the most practical existing machine.

Fig. 2 shows in dotted lines a replot of hardness penetration curves of the paper with various gas pressures and approximately 50 hours of duration of treatment. The paper does not state whether each curve

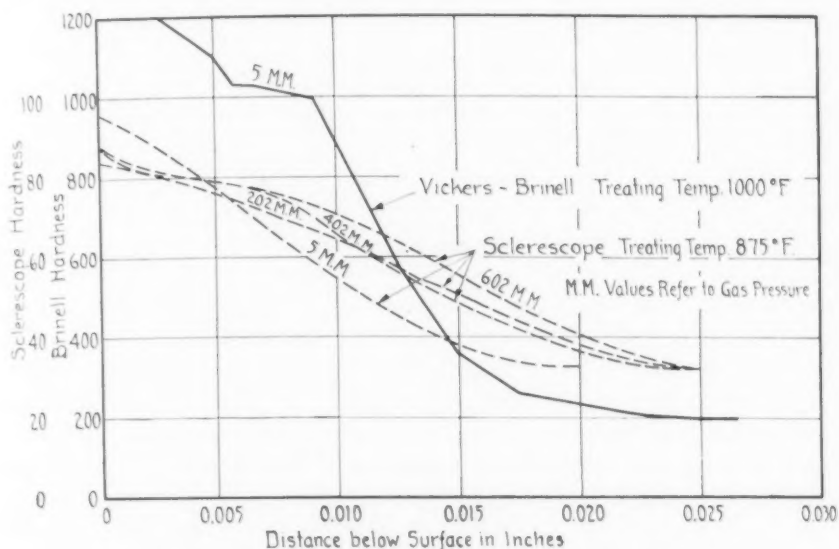


Fig. 2—Hardness-Penetration Curves of Nitrided Steel.

represents an average characteristic derived from several samples. This is an important point as we have found discrepancies between hardness penetration characteristics (of specimens taken from a single bar and subjected to supposedly identical treatments) greater than the discrepancies indicated in the four dotted curves in Fig. 2.

Fig. 3 shows an actual condition of this description where a 25-hour treatment was applied at 1000 degrees Fahr. and with about 5 millimeters gas pressure and we know of no reason for the spread of the curves. We are forced to conclude that there exist in the nitriding treatment variables which are at present entirely unrecognized. If this is a correct belief, it is evidently necessary that any conclusions be based upon foundations broad enough to represent average conditions.

We have added to Fig. 2 a curve showing the Brinell hardness penetration as determined on a single sample by the Vickers hardness tester. The analysis of the steel is the same as that referred to in the paper and the sample was nitrided for 50 hours at 1000 degrees Fahr. We are not prepared to defend the precise accuracy of these Vickers hardness values. It is, however, quite clear that the hardness at all points of the curve above the core is somewhat higher than the Vickers readings indicate. This is due to the fact that the Vickers machine must always read a little low on a surface which is backed up by softer material. The translation from Vickers to Brinell we neither guarantee nor condemn.

This translation was made by use of the Vickers conversion table. There is no attempt in the figure to harmonize the scale of the scleroscope and the Vickers (that is Brinell) readings. The two scales are shown in the figure. From the scleroscope-Brinell conversion table circulated by the Crucible Electric Steel Corp. the highest reading shown in the paper would

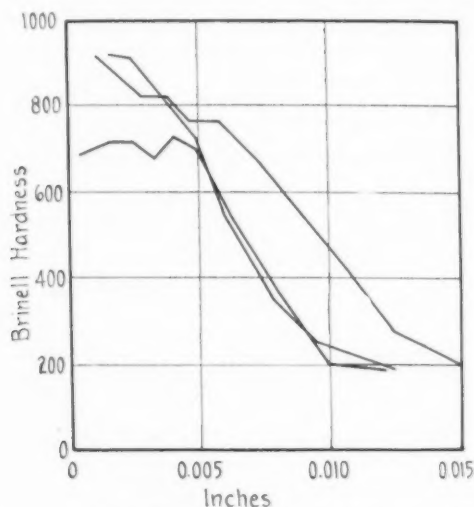


Fig. 3—Hardness-Penetration Curves of Nitrided Steels.

correspond to about 730 Brinell. From the conversion table of the new A. S. S. T. HANDBOOK this reading would be 712 Brinell. The Vickers machine showed 1210 Brinell at a point 0.0025 below the surface. All readings closer than this to the outer skin were so badly spalled, due to the brittleness of the material, that the impressions could not be measured. This gives rise to a very important question in connection with nitriding using gas pressures above atmospheric, namely, "Is the material near the surface utterly brittle, or does it verge toward a slight measure of ductility?"

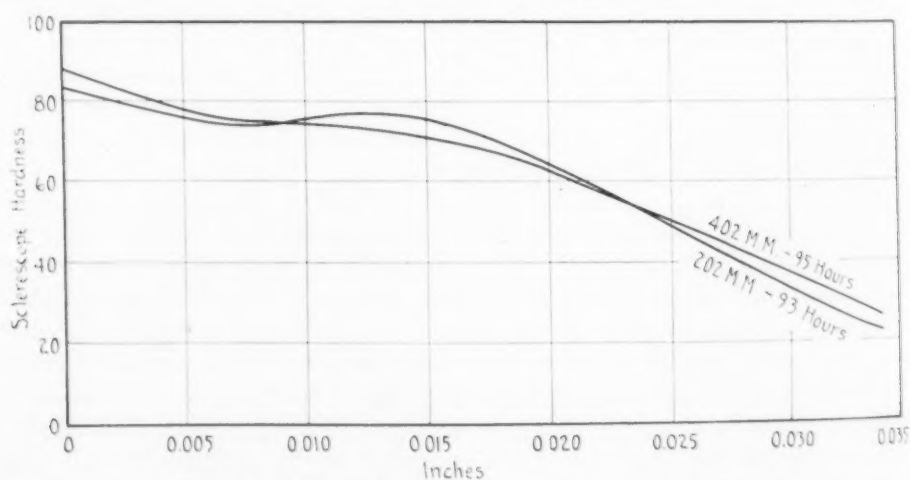


Fig. 4—Hardness-Penetration Curves of Nitrided Steel.

1929

The Vickers machine offers the only practical method we have found for answering this question. Those of us who have struggled with wear resisting problems are pretty thoroughly united in the belief that wear resistance is some composite function of hardness and ductility.

Referring further to the actual hardness gradient as compared to the gradient revealed by the scleroscope, it seems logical to believe that the steeper the hardness gradient, the greater will be the discrepancy between the scleroscope reading and the actual surface hardness. This means that if the gradients covered in the paper had been determined by the Vickers machine these gradients would have been steeper. Furthermore the curves shown with steep gradients in the paper would be steepened up to a greater extent than those showing a flatter gradient. There is, therefore, better ground than appears in the paper for believing that certain of the listed indicated results are real. Discussing these tabulated results in more detail, we can accept the first item because the steepness of hardness gradient is such that we have no confidence that the surface hardness as shown for the short time treatments is within several hundred points Brinell of the actual hardness. We have secured Vickers hardness on the outer surface of 1000 Brinell in 5 hours. It is entirely possible that under pressure a maximum hardness is achieved very quickly and that this then falls off with time, following even in the earliest stages the result referred to in item 2. This second result we think may be true though we consider the margin shown in the curves of the paper to be narrower than the variations of the scleroscope. With the steepening of the upper ends of the short time curves, which we expect to find by Vickers measurements, it is logical to believe that this item is justified.

Item 3 appears to be fairly well demonstrated, although there are rather outstanding exceptions to the law, in the curves shown in the paper. For example, Fig. 4 shows an almost identical case with 202 millimeter gas pressure and 93 hours treatment as compared with the case produced with 402 millimeters gas pressure and 95 hours treatment. Item 4 seems to us to be more nearly proven than any of the other results. We presume in item 5 that the limitation of depth within the general limits for nitrided cases is intended. We consider the study of the effect of gas pressure on nitrided products to be quite important. We suggest that in future work on this subject there are very simple ways for doing the preliminary work which enable the investigator to narrow down the range of his investigations rapidly and at comparatively low cost. For example, we recommend taper grinding through the case of the specimen instead of grinding off the entire case between readings as was carried out by the author of the paper. This is quicker, cheaper, and preserves a permanent record of all hardness measurements as well as furnishing the opportunity for repeating these measurements, if desired.

The following method was developed at Mellon Institute in Pittsburgh by Dr. J. H. Young and Dr. A. W. Coffman, with whom the Molybdenum Corp. of America are associated in research on the nitriding of steel.

When steel is nitrided in ammonia gas, a change of weight takes place.

This change of weight is usually a very good indication of the amount of nitrogen absorbed by the steel. There are sufficient apparent departures from this to take it out of the class of absolute natural law and it is essential to check all vital points of hardness, microscopic and other tests.

Fig. 5, Curve A, shows rate of change in the dissociation of nitrogen and hydrogen from the ammonia gas, as a function of temperature and with pressure constant at a point above atmospheric pressure by the amount necessary to carry the effluent gas out through the bubbler. This figure shows

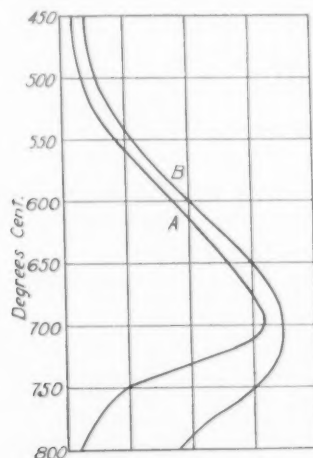


Fig. 5—Rate of Change of Ammonia Dissociation—Curve A. Gain in Weight During Nitriding—Curve B.

that the maximum rate of dissociation in a pyrex tube and with the particular steel employed occurs at 700 degrees Cent. We emphasize the type of steel because in a pyrex tube containing only the ammonia gas, practically no dissociation of the elements of the ammonia gas occurs at the usual nitriding temperatures. The steel apparently has a catalytic action assisting this dissociation when it is present in the heated ammonia gas. It seems possible that with radically different steels the point of dissociation may occur at some other temperature.

Fig. 5, Curve B, shows the nitrogen absorption of steel as a function of temperature and as determined by the increase in the weight of the specimen. The maximum rate of absorption of nitrogen occurs at 700 degrees Cent., which we have seen is the point of maximum change in the rate of dissociation. A similar procedure in connection with the study of the effect of pressure of gas seems to us to be a logical first move. Selected samples from these nitriding runs should then be examined for hardness, penetration, ductility, structure, etc. If the findings of the paper are borne out and it is demonstrated clearly that surface hardness is some inverse function of pressure, it would be quite logical to investigate a graduated pressure cycle, starting with a high pressure to secure quick and deep penetration and then dropping down to practically atmospheric pressure to secure the maximum surface hardness. Evidently there is room for a tremen-

dous amount of work to determine the most economic cycle, but in view of the great possibilities of nitrided steel products, any feature which offers logical hope of shortening the time necessary for the production of a given nitrided case, is worthy of extensive investigation.

Written Discussion: By Dr. V. O. Homerberg, Massachusetts Institute of Technology.

Any method of operation which will shorten the time period in the use of ammonia for the surface hardening of these special steels is most welcome. The work of the author is of sufficient importance to warrant further investigation. There are certain points in the present paper to which I should like to call attention.

The amount of combined nitrogen and the depth of the case produced are dependent on (1) temperature, (2) time of exposure and (3) the extent of the dissociation of the ammonia. The last variable is controlled by the regulation of flow of the ammonia gas and varies inversely as the pressure. No data are given which might indicate that the dissociation of the ammonia was observed. Bubbling the exit gas through water will not serve as an indicator since the water will soon become saturated with ammonia and the resulting bubbles may contain varying proportions of undissociated ammonia, nitrogen and hydrogen. I should suggest that the author conduct further experiments keeping the dissociation constant and allowing the change in pressure to be the only variable.

A question arises as to the character of the scleroscope impression. Any chipping of the case will give a low reading. I have obtained an equivalent Brinell hardness of 1000 on the surface as determined with the Herbert pendulum. The nitriding temperature was 875 degrees Fahr., the same as used by the author, and the dissociation was constant at 30 per cent. Several determinations on the same specimen by the Vickers hardness tester gave a constant reading of 960 Brinell. Obviously, such a high reading corresponds to a greater hardness than 90 to 100 scleroscope. I hope that the author will examine the scleroscope impressions in order to determine whether or not chipping took place in his tests. Also, I hope that a microscopic study will be made to detect any variations in the structure of the case with changes in pressure.

At the present time, I am of the opinion that the great hardness obtainable in this process is due to the formation of complex nitrides which are in dispersion corresponding possibly to "critical dispersion". The function of the aluminum is very much as stated. A steel containing this element along with chromium will maintain great surface hardness at elevated temperatures.

The author speaks of the ammonia treatment as a "nitration" process. Others have appended terms such as "nitrogenizing" and "nitrodizing". I have always used the term "nitriding" as the correct one, since it indicates that nitrides are formed in the operation.

Written Discussion: By P. A. E. Armstrong, consulting engineer, New York City.

Nitriding the surface layer of steels is a very old but quite interest-

ing process. Richard Hodgkinson, British Patent 19,493—1906, deals very completely with hardening the surface of steels by nitrogen derived from ammonia gas or compound yielding ammonia gas and uses acetylene as an additive gas. Also as much carbon as it is physically possible to get in, which is very little at low temperatures. Hodgkinson states that steel can be case hardened so that the surface is more than file hard even at such low temperatures as 300 degrees Cent. He also claims for his process the now well-known fact that at temperatures from 300 degrees Cent. and within the range now being used for nitriding, namely under 600 degrees Cent., that steel case hardened by this process will not warp, twist or move. Dr. Hodgkinson's work was conducted many years before Fry's of Krupp's, Essen, Germany.

Fry states that steels hardened by nitrogen should be heated to temperatures below the eutectoid point, and gives the eutectoid point at 580 degrees Cent. As a matter of fact the eutectoid point is somewhere about 608 degrees Cent. It is interesting to note here that Fry in his patent speaks of the peritectoidal point, but in his article appearing in the *Kruppsche Monatshefte* of September 1923, on page 147 he says, accurately translated: "It should be mentioned that in an earlier conception of the diagram in order to explain the double irregularity of the concentration-depth curves, a peritectoid transformation as 580 degrees was assumed. Later investigations showed that the transformation at 580 must be considered as the eutectoid transformation. The concentrations given in the diagram are only approximately correct due to the slight accuracy of the testing method used.

The lowering of the eutectoid point by nitrogen impregnation was well known prior to 1921. My experiments have shown that Dr. Hodgkinson's process patented in 1906, now expired, and not patented in the United States that I have been able to find, is more efficient than the later Fry process, which, in some measure, deals with nascent nitrogen derived from ammonia gas. The use of benzine gas with ammonia, which is now one of the principal methods employed for nitrogen surface hardening of steel, is very similar to the Hodgkinson method. Each of these gases, namely acetylene and benzine, is a hydrocarbon gas. Hodgkinson says that if you do not use enough hydrocarbon gas you are wasteful of ammonia gas and vice versa, and gives a proportion that works out very well. In this Hodgkinson is right and if benzine is to be employed larger quantities than are now customary should be used. If the temperatures are held below about 600 degrees Cent., or even somewhat higher, there will be no carbon impregnation that will be at all troublesome. When using ammonia gas alone, or with very small quantities of benzine which for all practical purposes may be disregarded, the ammonia gas breaks up in the heated chamber into its components hydrogen and nitrogen. Both of these gases play an important part in the process of nitrogen impregnation of the surface of steel or alloy steel. The ordinary process of case carburizing using hydrocarbon gases is well understood and there will be no dispute I think when I state the following fact; the surface layer of the steel is

impregnated with carbon and is decarbonized by the hydrogen so that a reversible process is going on all the time. There is, however, greater carburization than decarbonization, hence the surface of the steel can be carburized. A similar cycle takes place when ammonia gas is employed. The nascent nitrogen liberated from the ammonia combines with the alloy steel. The theory surrounding such combination is not important here. The nascent hydrogen does not immediately combine with the free nascent nitrogen from which it was separated, but combines with the nitrogen in solution in the surface of the steel being treated and denitrifies the surface layer of the steel. This process is not 100 per cent reversible in the same given time, hence nitrogen does remain in the steel. The surface layer will lose more of its nitrogen than the layer just under the surface. Hence where ammonia is used for nitriding the characteristic soft outer layer is present, and this soft layer is of greater magnitude than is generally appreciated because of the inadequate methods we have for testing this extremely thin and underlying hard layer. Nitriding by ammonia gas under pressure is speedier than the same operation carried on at substantially atmospheric pressure. The nitrogen penetrates quicker and the hydrogen denitrifies quicker. However, in the same given time the pressure system will result in a greater residual nitrogen content than when the pressure system is not used.

I have been actively conducting a series of experiments for a considerable time. The results of these experiments I do not feel at liberty to disclose in their entirety at this time. However, I have definitely proved to my satisfaction at least, that a catalyzer is a necessary factor in the successful surface impregnation of steels by nitrogen using ammonia gas, and that the hydrogen must be employed for other purposes than taking out the nitrogen content in the surface layers of the alloy steel being treated. This is readily accomplished by having available a substance that has a greater affinity for hydrogen than hydrogen has for nitrogen. By my process I have been able to cut down the time for nitriding by more than 50 per cent over the methods generally employed. Temperature is not an important factor except that if low temperatures are employed, that is below the eutectoid point, the alloy steel does not warp or twist. This was well disclosed by Hodgkinson and I suppose it will be conceded that many years before Hodgkinson it was generally known that tempering temperatures, that is under 500 degrees Cent., do not have a tendency to materially warp or twist a steel or ferrous article which is not under strain from some previous process. One of the principal advantages alloys have is that they will form very stable nitrogen compounds retaining the nitrogen in solution in the ferrous product and making denitrifying by hydrogen somewhat more difficult. The use of alloys for strengthening the core is not dependent upon any nitrogen treatment and was extremely well known to the art many years ago.

R. SERGESON: I would like to ask if anyone has done any of this high-temperature nitriding, that is, at 1000, and under this high pressure, around 600 millimeters. We have found that the results that Mr. Hobrock has ob-

tained we could obtain with about 1000 degrees temperature, the same results that he obtained at 875, under pressure, but I have never seen any nitrided cases that were greater than 0.035. I would like to know if anyone has used any different pressures to get a greater depth of case than that and carried the hardness in probably more on a straight line rather than to drop it down sharply. I would also like to ask if Mr. Hobrock's samples were heat treated. That is, he states there they were annealed at 1000 degrees, which, of course, is below the critical and would not cause any grain refinement. Generally, if you heat treat this material you get a better penetration in the heat treated state than you do in the as-rolled state.

A. C. JONES: Someone has raised the point about the percentage of alumina and aluminum in some of these nitrided steels. I would like to ask the question of someone who might be able to tell us, perhaps Professor Homerberg, what influence an excess amount of alumina in the steel has on the nature of the nitrided case, on the operation of nitriding, the speed of nitriding, and the hardness of the case.

DR. V. O. HOMERBERG: The only means that I know of determining whether or not there is any appreciable amount of aluminum oxide in the material is to examine it microscopically. In a previous discussion it was stated that the aluminum oxide might be present in such small particles that you were not able to see it under the microscope. I think, however, inasmuch as a lot of the specimens that I have looked at show very little aluminum oxide, and since the specimens after heat treatment give such very, very high values for impact, that I am justified in stating that I think practically all of that aluminum is present, possibly, in solid solution. I do not know anything about its influence under those conditions, and, therefore, I do not know anything about the influence of the aluminum oxide on the character of that nitrided case.

DR. O. E. HARDER: In the case of Mr. Hobrock's paper, I would like to know if he has made any determinations with reference to the character of this soft outer zone. In certain work which we have done, there seems to be formed on the outside a brittle, needle-like structure which is more or less porous, and I am wondering if this soft outer zone which he has observed is of that character.

R. H. HOBROCK: I have made no examination of that soft outer zone at all.

A. B. KINZEL: The matter of the use of the scleroscope in the determination of these curves has been mentioned by several of the members discussing this paper, but it cannot be overemphasized. All the conclusions are based on the hardness figures, and the hardness figures are determined by the scleroscope, a method which in the past has proved unreliable for quantitative work. I think that the least the author should do is to include in this paper a justification of that method.

The author also draws certain conclusions from the microstructure of nitrided specimens. In my experience, it is not possible to see any case structure in the microscope, without first reheating the specimens above what Fry calls the peritectoid temperature, and from the microstructures described

by Mr. Hobrock, I take it he did the same thing. Just what conclusions one may draw as to the condition of the material before heating by examination after it has been heated is rather difficult to say. I do not believe that such reasoning is particularly sound.

With regard to the general theory of the hardness of the case, the author states that it is probably another case of the precipitation of particles, and, along with Dr. Homerberg, I agree with him that this is a big part of the story. But, due to a number of anomalous conditions, particularly the fact that one cannot reproduce this hardness by reheating to just above the peritectoid temperature and requeenching, and drawing there is probably more to the mechanism than the simple matter of precipitation of particles.

Author's Reply to Discussion

Answering Mr. Eaton's questions, I do not know the grain size. This was a preliminary sort of experiment. The grain size was not determined. I used the scleroscope because that was the best of only three instruments that were available at Purdue, the Brinell hardness machine, the Rockwell machine, and the scleroscope, so I necessarily selected the scleroscope.

Another question by Mr. Eaton was about the curves, that is, whether or not they were average curves. They are not the result of a tremendous number of determinations. Some of the curves at the shorter periods I made from just one determination. The long-time curves were checked; a maximum of four samples to check those long-time curves and a minimum of three. The material that I used was all from one bar of steel.

Now, in regard to the great variations that Mr. Eaton found in some of his experiments, let me just suggest this, a thing that I thought about at the time, that I thought of this morning, and that I want to bring before you. That is, we do not know by chemical analysis whether the percentage of aluminum that we measure in the steel is metallic aluminum or whether it is there as an oxide, and from theoretical considerations it seems to me that that is very important. It seems to me that when we really get into this aluminum steel business, we will have to know exactly how much metallic aluminum is there. I should like sometime to do experiments with these alloys made in the vacuum furnace to avoid oxidation of the aluminum. This was a commercial alloy, and, as described in a paper this morning, a lot of them are made, of course, in the atmosphere, with a protecting slag, so we get aluminum oxide formed, some of which is included in the metal. Now, we may not even be able to see that aluminum oxide under the microscope, it may be there at such high dispersion that we cannot see it, and yet it may still greatly affect the properties of the metal. I want to mention that in connection with the first paper we had this morning by Mr. Breeler, I think.

Answering Dr. Homerberg, I did examine the impressions made by the scleroscope and found no chipping.

Dr. Homerberg also objected to the use of the test tube of water as a rough indication of the rate of flow of the gases on the ground that the water would absorb some of the gas. Of course the water does absorb some of the

gas when the water is freshly added, but it soon becomes saturated and may then be used as a flow indicator in a satisfactory way.

I made no examination of the exhaust gas to determine the percentage of decomposition of the ammonia. I do not think that the percentage of decomposition at the exhaust means much. The decomposition hardly goes on at all at this temperature without the presence of iron (or other catalyzers). Furthermore, I think that the reaction between the aluminum and the nitrogen requires active or nascent nitrogen since nitrogen in the molecular condition is notoriously inactive. I think then that the reaction resulting in the decomposition of the ammonia and in the formation of the aluminum-nitrogen compound goes on in the layer of absorbed gas—in places where a small amount of gas is in contact with a very great surface of the alloy. The amount of decomposition of the gas would then depend on the rate of flow of the gas and the amount of metallic surface available for the adsorption and the absorption of the gas—temperature and pressure remaining the same. In other words nothing is gained by having a large percentage of nitrogen and hydrogen gas on the retort containing the material to be treated—as a matter of fact such a condition might even be detrimental in that it (especially the hydrogen) might keep the ammonia gas from contact on the surface.

Mr. Eaton has considered the use of the scleroscope in detail in his paper, and has compared it with other types of hardness testers especially the Vickers. He concludes that my results might be even more marked had I used another type of hardness tester. Of course this is no place to discuss hardness, and I do not propose to do so. I appreciate the traditional unreliability of the scleroscope—yet I have found that when care is taken in keeping the tester clean, level and so forth, and when it is used in the examination of substances that do not have large crystals or heterogeneous surfaces, when the surfaces are clean and quite smooth—then I have had no difficulty and have been able to reproduce results. After all there is something to be said for a hardness tester to be used in this kind of work that does not penetrate the surface more than the scleroscope does. On the other hand there may be great advantages in the use of the Vickers—the matter of hardness testing has not been solved by any means. As I mentioned a while ago, I used the scleroscope because it was the best that I had available and because I could get reproducible results. I do not pretend that the theory concerning the hardness of these alloys is entirely correct—I propose it as a foundation upon which to build further research—it may be entirely wrong—but it seems to explain a good part of the observed phenomena. Mr. Kinzel objects on the ground that the hardness cannot be reproduced by heating above the peritectoid and quenching. Careful examination of the theoretical portion of my paper, which I did not have time to read, will show that one would not expect this to be possible—it would require the formation of a solid solution with the aluminum nitride as one constituent—I do not think this is possible. Heating to the peritectoid results in softening of the case because of the formation of interference particles that are too large.

DEOXIDATION OF STEEL WITH SILICON

BY C. H. HERTY, JR. AND G. R. FITTERER

Abstract

This paper presents an outline of the function of silicon in the basic open-hearth process, and the results of an investigation on the mechanism of deoxidation of steel with silicon. The formation and elimination of silicates is briefly discussed.

As a result of the experimental work described in this paper it may be concluded that the mechanism of deoxidation with silicon is first, the formation of silica by the reversible reaction of silicon with FeO and second, a fluxing of more FeO by the silica, forming a ferrous silicate inclusion. Complete deoxidation of steel is impossible unless the oxide formed is absolutely infusible and insoluble at steel-making temperatures. It may be also concluded that silicates high in iron or manganese will be more readily eliminated from the bath than high silica particles, due to their lower melting point, with consequent increased rate of coalescence.

FUNCTION OF SILICON IN THE BASIC OPEN-HEARTH PROCESS

Silicon in the Charge

SILICON is introduced into the basic open-hearth furnace under two conditions: first, in the charge, and second, as a deoxidizer at the end of the heat. Most of the silicon in the charge enters through the medium of the hot metal or cold pig iron. The silicon serves three important purposes during the first part of an open-hearth heat, as follows:

1. Its oxidation by means of iron oxide in the open-hearth slag is attended by a large evolution of heat and the temperature of the bath is raised considerably by this oxidation. The temper-

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A paper presented before the tenth annual convention of the society held in Philadelphia, October 8 to 12, 1928. Of the authors, who are members of the society, C. H. Herty, Jr., is physical chemist, and G. R. Fitterer is junior metallurgist with the Pittsburgh Experiment Station, Bureau of Mines, Pittsburgh. Manuscript received July 9, 1928.

ature of the hot metal added to the furnace is usually about 2350 to 2400 degrees Fahr. (1288-1324 degrees Cent.). The oxidation of the silicon from the charge will raise the temperature of the bath about 100 to 150 degrees Fahr. (38-65 degrees Cent.) when the hot metal contains 1 per cent silicon. This rise in temperature is very important in that it speeds up the melting of the scrap in the furnace and helps to shape up the slag more rapidly than if the heat requirement were met entirely from the fuel.

2. The second function of silicon is to form a fusible slag with the lime charged and with the other oxides such as manganese and iron oxide formed or added during the process. If the silicon in the charge is too low the slags will be viscous throughout the heat, unless excessive amounts of other fluxing agents such as fluor spar or iron oxide are added. If the silicon in the iron is too high there is considerable danger of very thin slag being formed and persisting throughout the heat. Furthermore, excessive erosion of the furnace lining will take place as the slag will be siliceous and very fluid, and will react rapidly on the dolomite banks and bottom of the furnace. The proper silicon content of the charge will vary for different practices, depending on (a) the type of steel being made—that is, low carbon or high carbon; (b) the amount of limestone charged; and (c) the amount of manganese in the charge.

3. The third function of silicon is to deoxidize to a certain extent the slag formed during the melting of the scrap. If the silicon content of the iron is too low, excessive oxidation of the slag is likely to persist throughout the heat; and if the silicon content is too high, the slag may be so thoroughly deoxidized that it will be difficult to dissolve the lime without large additions of iron ore or spar.

ELIMINATION OF SILICON FROM PIG IRON

The elimination of silicon during the early period of an open-hearth heat is usually very rapid. The only condition to be fulfilled for this rapid elimination is that there be iron oxide present for the oxidation of the silicon. If the amount of iron oxide formed during the melting of the scrap is small enough so that the silicon in the charge reduces it to a low value before all of the silicon has been eliminated, the silicon elimination will necessarily be slow.

On the other hand, if there is a large amount of iron oxide present when the hot metal is added, silicon elimination will be extremely rapid and there will be only traces of silicon left in the bath long before the whole charge is melted.

Fig. 1 shows the elimination of silicon¹ in two heats—A and L. In this figure the per cent silicon in the metal and the per cent iron oxide in the slag are plotted against time. In heat A,

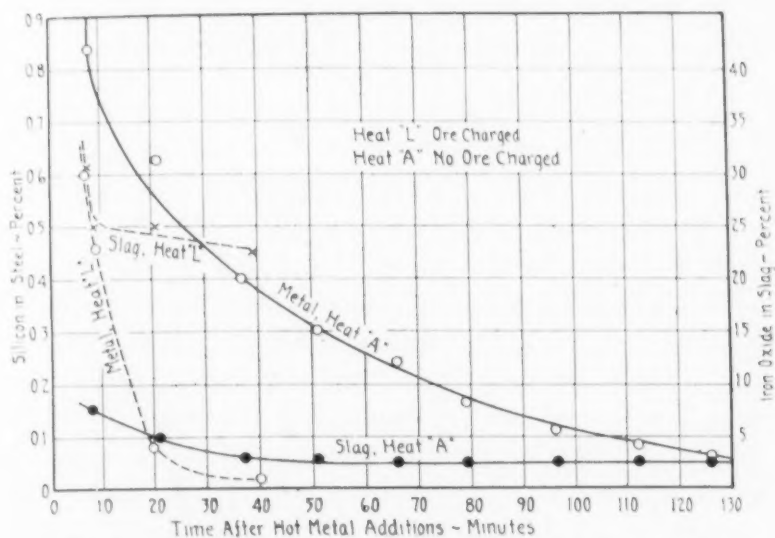


Fig. 1—Elimination of Silicon in Basic Open-Hearth Practice.

heavy scrap only was charged and the amount of iron oxide in the first slag was fairly low; likewise the total amount of slag formed at this time was low. In heat L, ore was charged with the scrap and the iron oxide content of the slag was high when the hot metal was added. In these two heats we have the two extremes of silicon elimination—heat A showing slow elimination because the slag was low in iron oxide, and heat L showing a rapid elimination due to the high iron oxide content of the slag.

SILICON ADDED AS A DEOXIDIZER

In discussing silicon additions at the end of an open-hearth heat it is necessary to define two types of steel: (1) rimmed steel, where no silicon is added as a final addition; and (2) killed steel, where silicon is almost always added.

¹Keats, J. L., and Herty, C. H., Jr., "Elimination of Metalloids in the Basic Open-Hearth Process," *Transactions, American Institute of Mining and Metallurgical Engineers*, Vol. 73, 1926, p. 1079.

In order to avoid confusion in differentiating between various types of inclusions the following nomenclature has been adopted:

Ferrous silicate—Inclusions containing iron oxide and silica in any proportions.

Manganous silicate—Inclusions containing manganese oxide and silica in any proportions.

Iron-manganese silicate—Inclusions containing iron oxide, manganese oxide, and silica in any proportions.

Fayalite—Ferrous ortho-silicate of the composition $2\text{FeO}.\text{SiO}_2$.

Rimmed Steel

In making rimmed steels it is sometimes customary to reboil heats with silicon pig, a pig iron containing either 10 or 15 per cent silicon. It is generally considered that after a silicon pig reboil the metal will be lower in iron oxide than before the reboil. It is certainly true that when the silicon is added to the bath a certain amount of deoxidation takes place, but it is also equally true that in order to have a reboil occur—that is to say, for carbon elimination to recommence—iron oxide must diffuse from the slag into the metal in order to eliminate the carbon. It is entirely possible that oxides suspended in the metal before the silicon addition may be fluxed out of the metal by the fine mass of silica formed on the addition, and that the increased temperature resulting from the elimination of the added silicon may be sufficient to melt oxides which were infusible at the temperature of the bath before the silicon addition. Regardless of the final results, it is certainly true that when silicon is added for reboiling, ferrous silicates are formed in the bath and these must be eliminated if clean steel is to be made. The silicates so formed float out of the bath either as high silica or high iron ferrous silicates, the type of particle depending on the amount of iron oxide present to flux with the silica formed.

Killed Steel

In making killed steels, silicon may be added in two ways: first, the heat may be killed in the furnace with silicon and an extra addition of silicon made in the ladle; or, the heat may be killed in the ladle alone. In killing the heat in the furnace, enough silicon is added to quiet the bath and then at a time usually de-

terminated by the melter, the heat is tapped with the bath still dead and the extra silicon is added in the ladle to complete deoxidation as far as desired. Silicon is added to the furnace whenever the practice of recarburizing with pig iron is carried out. In this practice, if a high carbon steel is to be made it is very often the custom to take the heat down to a low carbon, and then add sufficient hot metal to bring the carbon to the desired specification. In this pig iron there is of course a certain amount of silicon, and the heat may or may not be killed at the time of the hot-metal addition. This depends upon the amount of silicon in the iron, the amount of iron added, and the iron oxide content of the steel at the time of the addition. It may be pointed out here that the iron oxide content of the metal depends primarily upon the carbon content and the temperature. In making some high carbon steels the bath may be reboiled with silicon pig. The same considerations hold here as in reboiling rimming steels, except that the ferrous silicates formed will always be high in silica, due to the low iron oxide content of high carbon steels.

In killing the steel in the ladle, enough 50-per cent ferrosilicon is added to ensure that the steel comes within the silicon specifications. In order to meet specifications it is necessary to add more silicon than specifications call for, inasmuch as a certain amount of silicon is lost through reaction with the iron oxide dissolved in the metal, and with iron oxide which diffuses into the metal from the ladle slag. The higher the carbon at which the heat is tapped the less silicon will be lost on the addition. At the same time that the silicon is added in the ladle, ferromanganese is also added, and in this case manganous silicates or iron-manganese silicates will be formed. These depend upon the amounts of the ferrosilicon and ferromanganese additions and on the amount of iron oxide present in the metal at the time of the addition. The elimination of these silicates from the steel is primarily a function of their fusibility. In general, the more fusible the silicate the more rapidly the particles will coalesce, and the larger the particle the faster it will rise to the top of the metal and be absorbed by the slag.

In general then, we may have three types of silicates formed in most steels: (1) ferrous silicates and iron-manganese silicates in rimming steels; (2) manganous silicates in killed steels; and (3) iron-manganese silicates in killed or semi-killed steels. A fourth

type of silicate, a particle very high in silica, is that which enters the charge in the pig iron or the scrap and is not eliminated during the heat. That this type of silicate is present is a matter of experimental fact² and must be considered in determining how clean the steel will be under any conditions. However, in this paper, which deals primarily with deoxidation with silicon, it will not be referred to again.

EXPERIMENTAL PROCEDURE

In order to obtain data on deoxidation with silicon, it was thought best to work with low carbon steel where the amount of iron oxide present in the steel could be determined with accuracy and where, by varying the amount of silicon added, different types of ferrous silicates would be formed. Accordingly, heats were made in an electric melting furnace in which 250 to 300 pounds of boiler punchings, plus a small amount of coke, were melted down and the melt "ored down" to 0.03 to 0.05 per cent carbon. Slag was formed by adding lime and iron oxide during melting, and the finishing slags were more or less typical of low carbon basic open-hearth steel slags, except that the manganese content of the slag was low. An average analysis of the finishing slag was—

Constituent	Per Cent	Constituent	Per Cent
SiO ₂	9.0	MnO	3.0
P ₂ O ₅	0.8	CaO	42.0
FeO	27.5	MgO	4.0
Fe ₂ O ₃	11.0	Total S	0.48

When the heat was ready to be tapped, 20 to 25-pound ladles of the metal were poured and varying amounts of 50-per cent ferro-silicon were added to the ladles. As soon as possible after the addition, the metal was poured into 3½-inch ingot molds and the action in the mold was carefully watched. One ingot was poured with no silicon added in the ladle, and from this ingot the amount of iron oxide dissolved in the metal was obtained by analysis by the Ledebur method—that is, reduction with hydrogen. The ingots were split longitudinally and sections were taken out at various points for the determination of ferrous silicates and for microscopic examination. The determination of silicates was carried out by the

²Herty, C. H., Jr., and Gaines, J. M., Jr., "Unreduced Oxides in Pig Iron and Their Effect on Open-Hearth Steel." (Manuscript report.)

Dickenson method. This consists of solution of the steel in cold 10 per cent nitric acid, with a treatment of the residues obtained with dilute hydrochloric acid for the solution of iron salts, and treatment with caustic to remove the silicic acid formed from the solid solution silicon in the steel. Sections of all ingots were milled and analyzed for iron oxide by the Ledebur method.

EXPERIMENTAL RESULTS

Type of Inclusions Formed

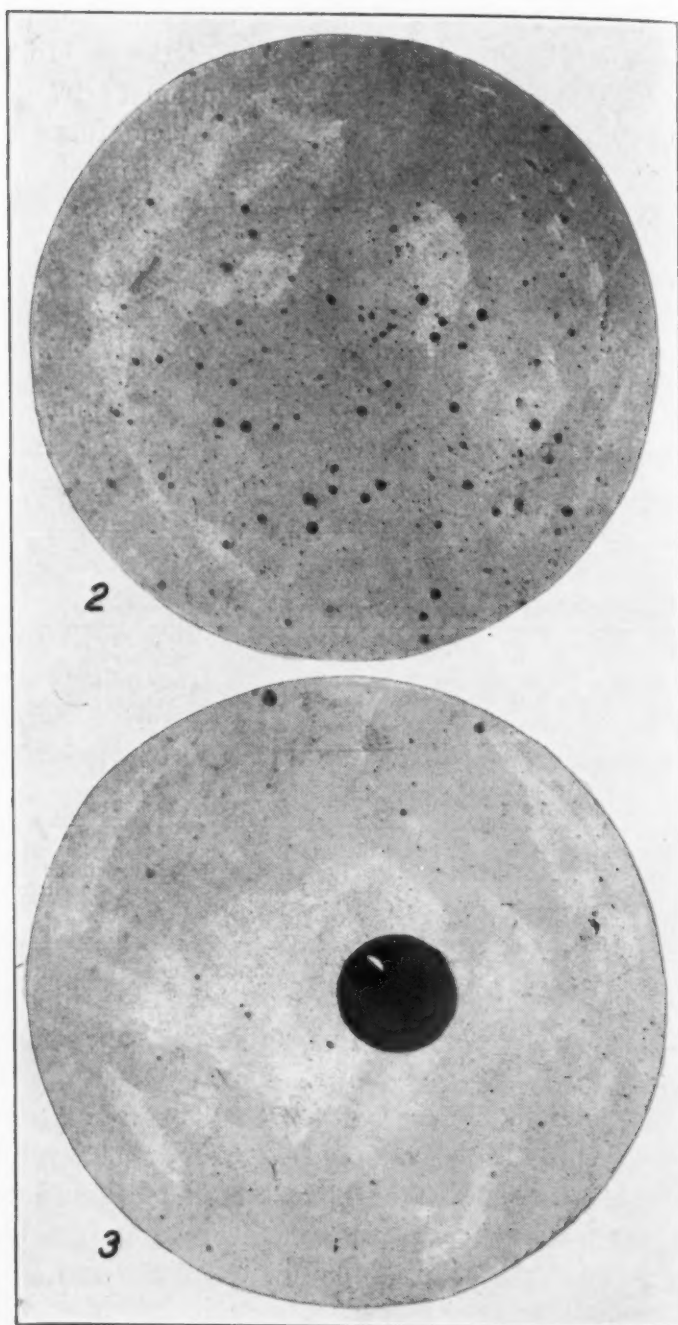
Fig. 2 and Table I show the types of inclusions formed on various additions of silicon, analysis of the steels, and approximate analysis of the inclusions.

Table I
Results of Deoxidation on Four Ingots

Ingot No.	Fig. No.	Per Cent Si added	C, Per Cent	Analysis of Steel		Approximate Analysis of of Inclusion,	
				Total Si, Per Cent*	FeO expressed as Oxygen, Per Cent	FeO	SiO ₂
1	2	None	0.046	0.004	0.075	100	0
2	3	0.027	0.048	0.005	0.052	90	10
6	4	0.132	0.069	0.050	0.021	35	65
9	5	0.330	0.048	0.212	0.011	5	95

*Some silicon was mechanically lost on all additions, and on ingots 2 and 6 some of the silicates rose out of the steel before the metal solidified. The silicon is expressed as total silicon. Some of this was present as silica and in the ingots containing high iron oxide the two cannot be differentiated by the Dickenson method.

Fig. 2 shows the iron oxide in ingot No. 1, to which no silicon was added. This metal contained 0.075 per cent oxygen (0.349 per cent FeO). Fig. 3 shows a large ferrous silicate particle surrounded by small spots of iron oxide. The silicon addition lowered the oxygen content of the steel from 0.075 to 0.052 and formed high iron ferrous silicates which, on account of their low melting point, were very fluid at the steel temperature and coalesced rapidly, giving large particles of iron silicate. Fig. 4 shows two types of inclusions—first, a large ferrous silicate inclusion with a duplex structure, and second, a glassy inclusion. The steel in which these inclusions were found was in the range between rimming and semi-killed steel. The oxygen content of the metal had been lowered to 0.021 per cent and the inclusions analyzed about 65 per cent silica. This average represents both types, and it is probable that the in-



Figs. 2 and 3—Photomicrographs Showing Inclusions in Ingots Numbers 1 and 2 Respectively. $\times 300$.

clusion with the duplex structure is somewhat lower in silica and the glassy inclusion is somewhat higher in silica than this average.

This inclusion has been tentatively given the name "intermediate type" to distinguish it from either the high iron ferrous silicate type or the glassy type, and is the result of formation of silica with a subsequent fluxing of a small amount of FeO from the metal. Fig. 5 shows the glassy type of ferrous silicate found in dead-killed steels. The oxygen content of this steel was 0.011 per cent and the inclusion analyzed approximately 95 per cent silica.

Thus, as increasing amounts of silicon were added to the metal the oxygen content of the metal was steadily decreased, and the type of inclusion changed from the very fusible high iron ferrous silicate through a ferrous silicate very high in silica with a definite duplex structure, to a siliceous inclusion which is very hard and brittle at ordinary rolling temperatures.

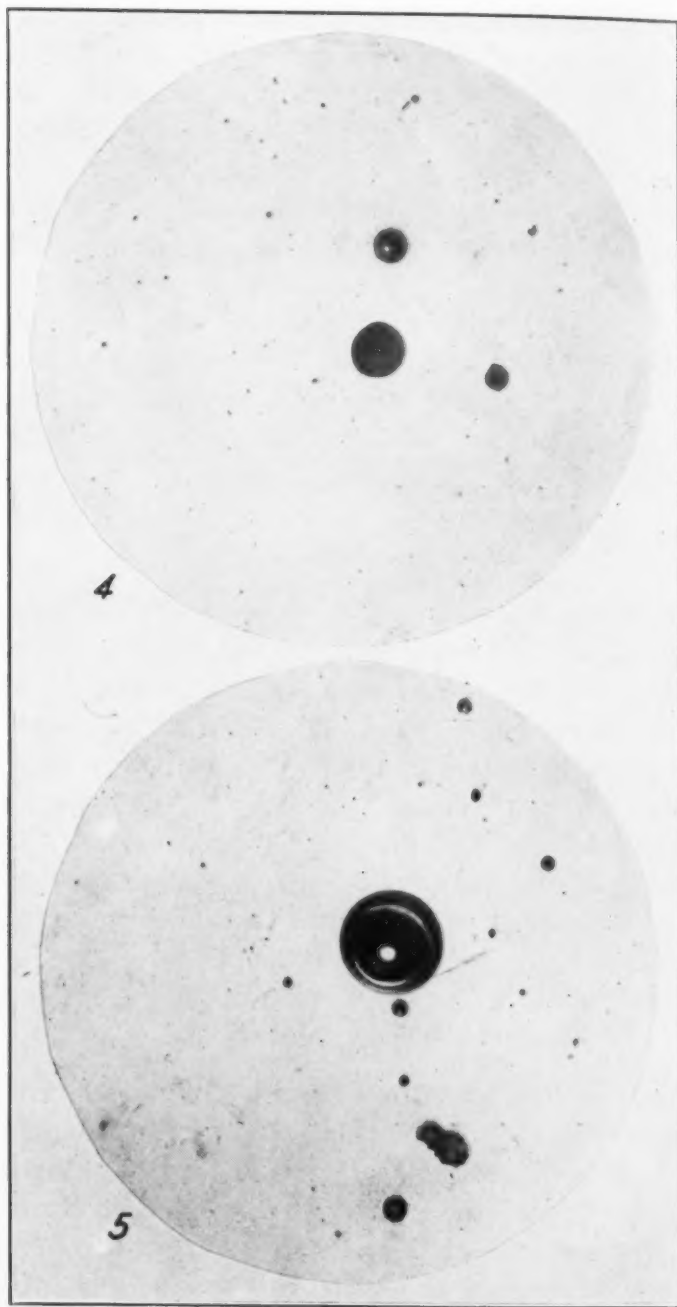
Segregation of Silicates

Ingots No. 6 and 9 were split longitudinally and samples were taken as shown in the photographs of Fig 6. The amount of silicates present was determined in each sample and adjacent samples used for microscopic inspection. The killed ingot (No. 9) showed a slight segregation at the center, with a maximum at the lower section, and the rimmed ingot (No. 6) showed a high segregation along the central axis and high segregation both at the top and bottom of the ingot. The segregation on these two ingots is typical of industrial ingots, with particular reference to the high segregation at the bottom of the ingot.³

Mechanism of Deoxidation

The mechanism of deoxidation with silicon is readily determined by a consideration of the types of inclusions found on ingots No. 6 and 9, as they are shown in Fig. 7. In the killed ingot all of the particles are of the glassy type. In ingot No. 6, on the other hand, the particles at the edge of the ingot are of the glassy type and in the center of the ingot they are of the intermediate type; the latter contains much more iron oxide than the former. The time between the silicon addition and the pouring of the metal into the mold was approximately 30 seconds, and in this size of ingot 1½ to 2½ minutes are required for complete solidifi-

³Dickenson, J. H. S., "A Note on the Distribution of Silicates in Steel Ingots." *Journal, Iron and Steel Institute*, Vol. 113, 1926, No. 1, pp. 177-196.



Figs. 4 and 5—Photomicrographs Showing Inclusions in Ingots Numbers 6 and 9 Respectively. $\times 300$.

cation. The steel at the edge of No. 6 would contain the types of inclusions found in the ladle to which the ferrosilicon had been added. At the time of the pouring of the steel there were evidently

only glassy ferrous silicates suspended in the steel. As solidification proceeded, these glassy particles absorbed iron oxide which was dissolved in the steel. The first action of silicon is therefore to form SiO_2 by the reaction



and the SiO_2 thus formed fluxes dissolved FeO out of the steel to form a ferrous silicate.

It should be noted here that the inclusions contained a small amount of MnO , which was evidently present as dissolved MnO in the steel before the addition of silicon.

The fluxing action of silica is of great importance in steel-making, inasmuch as a great deal of deoxidation of steel can take place through the fluxing out of MnO and FeO by silica. As the ferrous and manganous silicates are insoluble they tend to rise out of the steel, and by this double action—that is, fluxing and elimination through insolubility—the steel is cleaned. The amount of fluxing which will be done through a given silicon addition largely depends upon the amount of silica formed, the amount of FeO left unreduced, and the temperature. The silicate particles in the steel simply act as a slag phase and equilibrium tends to be set up between this slag phase and the melt. FeO will be fluxed out of the steel into the inclusions until a distribution ratio

$$\frac{\text{free FeO in the inclusion}}{\text{FeO dissolved in the steel}}$$

is satisfied. Were it not for the fact that in ferrous silicates a great deal of the FeO is combined as fayalite, it would be easy to predict the amount of fluxing which could be done by a given silica particle. However, the dissociation of fayalite into silica and FeO has not been quantitatively determined, and until this dissociation constant is known, the amount of free FeO in any given ferrous silicate inclusion must remain an unknown quantity.⁴

It is evident from the above that complete deoxidation of steel by fluxing cannot be accomplished, for as long as there is FeO present in an inclusion there will be some free FeO present, and since there is free FeO present in the inclusion there will be dissolved FeO in the steel. The only condition under which complete re-

⁴Experimental work on the dissociation of ferrous silicates is being conducted at the present time at the U. S. Bureau of Mines Experiment Station, Pittsburgh, Pa.

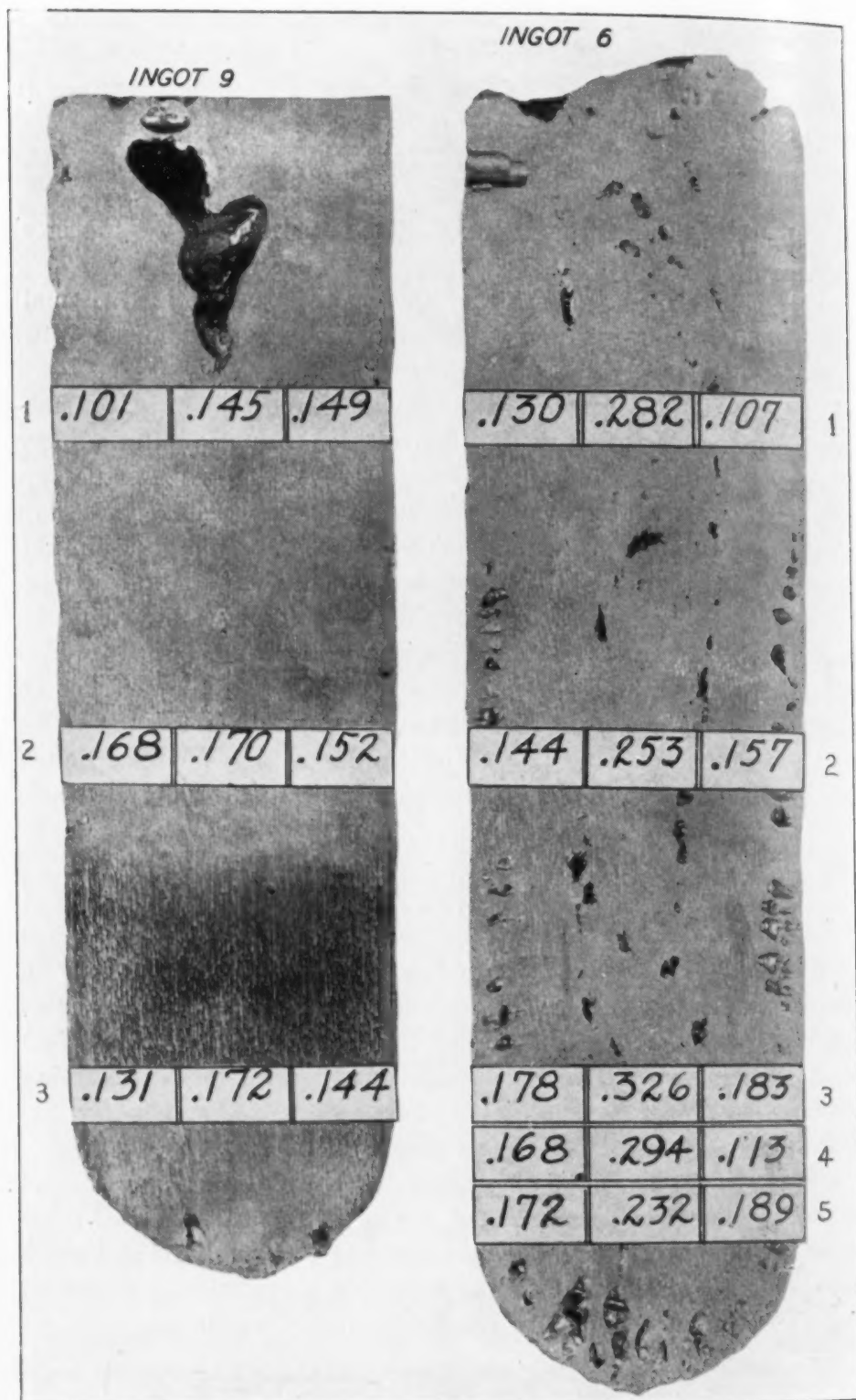


Fig. 6—Photograph of Ingots Numbers 6 and 9 Which Were Split Longitudinally and Samples Taken to Determine the Amount of Silicon Present. The Figures Indicate the Silicon Content. Ingot Number 9 is Killed Steel and Ingot Number 6 is Rimmed Steel.

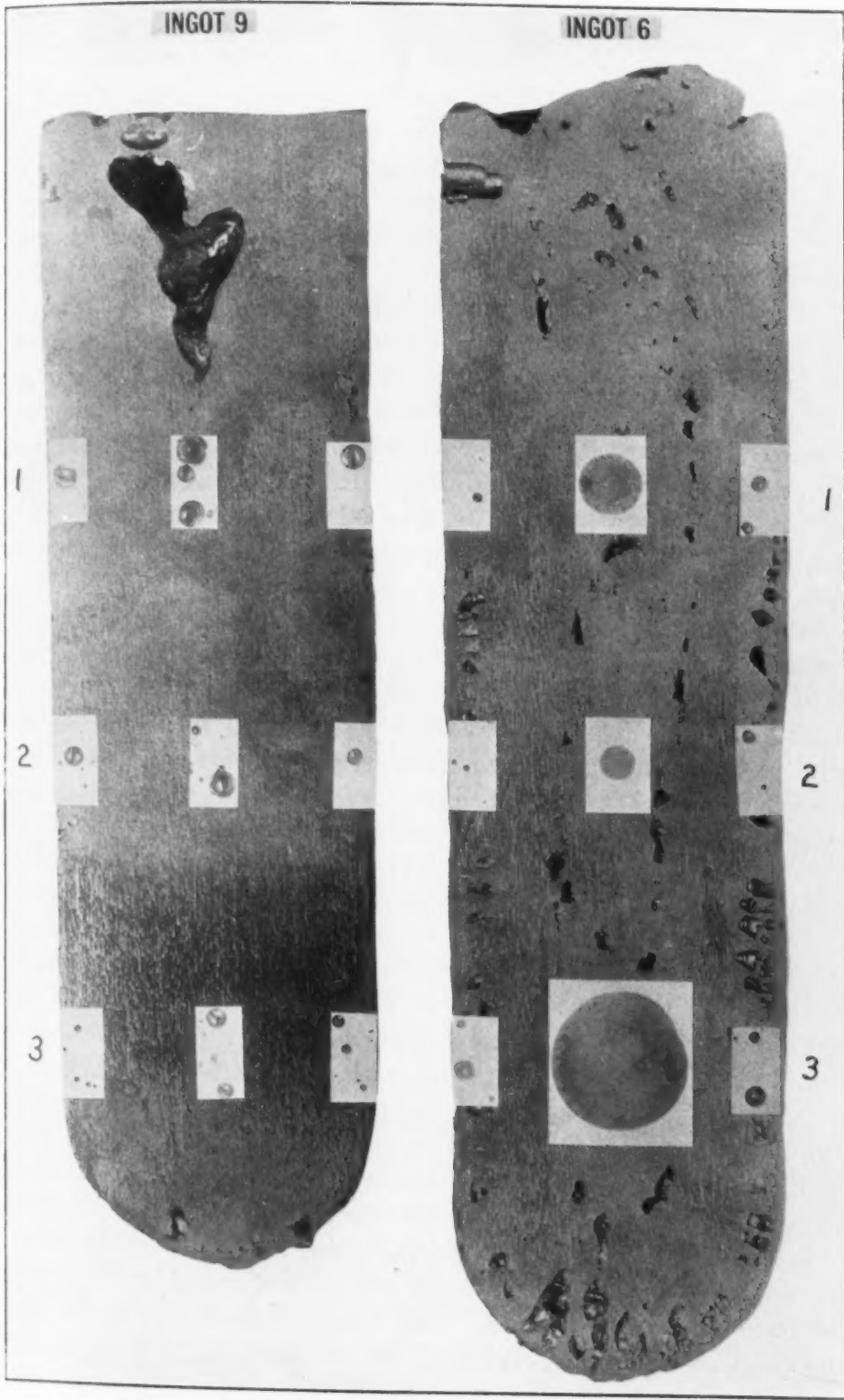


Fig. 7—Photograph of Ingots Numbers 6 and 9. Same Split Ingots as Shown in Fig. 6. Showing Photomicrographs Taken at Areas Corresponding to Those from Which Silicon Determinations Were Made. Ingot Number 9 is Killed Steel and Ingot Number 6 is Rimmed Steel.

duction of iron oxide could take place would be for an absolutely infusible and insoluble material to be formed on deoxidation.

EFFECT OF SILICATES IN STEEL

There are innumerable references in the literature to the effects of oxides in steel. Consumers are becoming more and more strict on specifications as to the cleanliness of steel. In general, the known and presumed effects of oxides in steel are: (1) they decrease the transverse strength; (2) they decrease the impact strength; (3) they increase the tendency for fatigue failure; (4) they increase the tendency for corrosion; (5) they cause surface defects in highly finished material; (6) they are contributory causes to butt cracks in ingots; and (7) they cause red-shortness in rolling under certain conditions. Unquestionably certain types of inclusions are more harmful than others, and "keying" by very small inclusions may reduce the tendency to slip, thereby increasing the tensile strength and decreasing ductility. The type of inclusion shown in Fig. 3 is almost identical with that found in wrought iron and cannot therefore be considered harmful for certain classes of work. On the other hand, the type of inclusion found in Fig. 5 is a very refractory type which will elongate with difficulty on rolling and certainly cause red-shortness if present in sufficient quantity. This type of inclusion would also seriously affect the impact strength of steel.

CONDITIONS FOR THE ELIMINATION OF INCLUSIONS IN STEEL

The rate of elimination of inclusions in steel is affected by particle size and specific gravity of the particle, viscosity of the steel, and mechanical agitation of the steel bath. The rate of elimination of inclusions due to separation by difference in specific gravity is increased by increasing particle size and by decreasing specific gravity of particle. We would thus expect a very light inclusion, which coalesced sufficiently to give a very large particle, to rise much faster out of the steel than a dense particle of a material which had such a high melting point that the particles would not coalesce. There would be, of course, any number of cases in between these two extremes. The two important factors are particle size and specific gravity of the particle. It is obvious that

small particles will not coalesce to give large particles unless the particles are fluid. In inclusions having a melting point lower than the temperature carried in the open-hearth furnace the particles will tend to coalesce, and in general the more fluid particles will coalesce more rapidly than the less fluid. An example of this is shown in Fig. 7, where the inclusions in the rimming steel ingot, having a low melting point, grew to a very large size, whereas the glassy particles which were probably pasty, grew only slightly. Agitation of the bath assists in elimination in that the particles are mechanically carried to the slag surface instead of having to rise through the steel, simply due to a difference in specific gravity between the particle and the steel. The one other condition necessary for elimination is that slag conditions be such that the particle will be readily absorbed when it reaches the slag-metal surface. The essential conditions for absorption are that the slag "wet" the particle and have an equal or greater surface tension than the particle. As the ratio

$$\frac{\text{surface tension of slag}}{\text{surface tension of particle}}$$

increases, absorption will be more rapid and at very low values of this ratio—that is, when the slag has a low and the particle a high surface tension—very little absorption would be expected. If the particles are not absorbed by the slag they will be thrown back into the metal when the heat is tapped. Unfortunately, practically nothing is known of the surface tension properties of different types of slags, but it is fairly safe to predict that the surface tension will vary considerably with the normal changes in chemical composition found in finishing open-hearth slags.

CONCLUSIONS

As a result of the experimental work described in this paper it may be concluded that the mechanism of deoxidation with silicon is, first, the formation of silica by the reaction



and second, a fluxing of more FeO by the silica, forming a ferrous silicate inclusion. Complete deoxidation of steel is impossible unless the oxide formed is absolutely infusible and insoluble at steel-

making temperatures. It may also be concluded that silicates high in iron or manganese will be more readily eliminated from the bath than high silica particles, due to their lower melting point, with consequent increased rate of coalescence.

DISCUSSION

DR. W. H. HATFIELD: Mr. Chairman and gentlemen: I rise to discuss this paper, if only to say how much we in England are watching the magnificent research work which is being carried through by Dr. Herty. The subject that he is dealing with is, of course, one of primary importance to all engaged in manufacturing steel, and we on our side are dealing with the same problem. In fact, the ingot committee of the Iron and Steel Institute are taking a very detailed interest in this subject, and therefore we particularly watch everything that Dr. Herty issues.

Now, Dr. Herty would be the last to expect us to take all his conclusions and accept them. We, naturally, knowing the high standard of his work, accept all his facts. It is the interpretation that he puts upon them that he must not be surprised if there is a little hesitation in entirely accepting.

For instance, Dr. Herty must be the first to acknowledge that when the gases come off liquid steel, there is a tremendous amount of hydrogen there. Now, surely a strongly reducing element like hydrogen must, indeed, play its part in the reactions which he is studying at the moment, and I do not know whether he mentions hydrogen in the paper, but I should like to hear from him this morning as to what his view is concerning the part played by hydrogen. The acceptance of Dr. Herty's views would be much easier if it were not for the disturbing presence of that highly reducing element.

Again, Dr. Herty is putting forward the suggestion that silica, SiO_2 , is soluble in steel. Now, from a purely theoretical point of view, one cannot take exception to his suggestion. Nevertheless, I submit to him that it is a very important deduction for him to make and I should be very much interested to know to what extent he actually has experimental evidence which will give him any indication at all that silica is in solution in liquid steel,—experimental evidence. Now, I say "silica" as distinct from "silicon."

Then, there is another very interesting phase of this matter which must be borne in mind and that is that these very excellent, constructive synthetic experiments which Dr. Herty is conducting are, as he told us, based on a very low carbon content, and we must always remember that in the open-hearth furnace, particularly in making a lot of the, I do not know whether you call them special steels in America, but steels of higher carbon content, the presence of the high carbon content in the steel must certainly have a bearing, an influence, on the solubility of the oxygen in the bath.

Now, I do not see that so far Dr. Herty has suggested the extent to which, if any, the presence of carbon will influence the solubility of oxygen in the steel. It may surprise Dr. Herty to hear me say, for instance, that I still have a very great doubt in my mind as to whether there is any consid-

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erable—and I use “considerable” in the sense that Dr. Herty would use it—solubility of oxygen in the bath when the carbon is at a high level. I have been doing a lot of experiments over the last year or two which lead me to doubt very much whether, in the presence of carbon as distinct from in these almost carbonless irons, the oxygen content is as high as we are wont to believe at the moment.

Now, it may interest Dr. Herty to know that Professor Desch, whom we met over here some time ago in Pittsburgh, has been doing very extremely carefully the oxygen determinations in some of our steels in Sheffield, and his more recent results show the oxygen content, the total oxygen content, at around about 0.005 per cent—acid open-hearth—which is very well below what one would anticipate from your figures. But, at any rate, I would be very glad to hear what you have to say about that.

Then, as regards inclusions and the separation of inclusions from steel, Dickenson's method is largely used, but I submit to Dr. Herty that if he reads the discussion on Mr. Dickenson's paper, he will find a very interesting contribution from Mr. Kalkloff. Now, Mr. Kalkloff took a sample of steel, separated the inclusions by Dickenson's method, and obtained a certain value. He then separated the inclusions by using sulphuric acid instead of nitric acid and obtained a much higher yield of inclusions. Therefore, it would certainly appear that in determining the silica content which is fixed in the inclusions and the silica which is in the steel, due recognition should be taken of the fact that it is extremely doubtful whether the Dickenson method does completely return all of the silica.

There is only one more point that I would make and that is this, which will give you all a lot of hope on this inclusion question.

If you properly refine steel in the electric furnace, as you may, you will reduce the oxygen content to an extremely low level, and, if you are successful, you will obtain steel under the best conditions which is virtually free from inclusions. That is the latest information, in a way, from Sheffield. It is only information yet, because while we are very proud of our electric steels that we are making and our acid open-hearth, doing the best we can, we still feel that with the knowledge which is now becoming available, steel will be made which is almost free from inclusions, and the avenue through which that will best be done will be the electric furnace.

A. L. FEILD: Mr. Chairman, there is one fact which seems to me might be emphasized to advantage in connection with this excellent paper and that has to do with Table I. You will notice there that steel with 0.21 per cent silicon is shown as having a glassy type of inclusion, 95 per cent silica and 5 per cent FeO. Now, as a matter of fact, that is not true in the case of steel which is first deoxidized with ferromanganese in the furnace before adding silicon.

Dr. Herty has already brought out adequately, it seems to me, the reactions which go on in the case of a furnace addition. I believe it frequently is the case where no furnace addition is made but where ferromanganese is put in the furnace, followed by ferrosilicon, that you can get a steel with 0.21 per cent silicon in which the silicates are not so glassy, and correspond to a

manganese silicate rather than an iron silicate. This fact, it seems to me, ought to be stressed a great deal. Otherwise, we might conceivably consider all steels of the normal type of killed steel, with 0.20 to 0.25 per cent silicon, as being more or less red-short.

A. T. CAPE: Mr. Chairman, Dr. Herty and gentlemen: May I bring up with Dr. Herty a matter of technique? In his investigation of the problem of inclusions I was wondering if any of the samples were examined by means of the petrographic microscope. It has been possible to examine the larger inclusions found in steel in thin sections and we have not been able to identify any of the inclusions as silicates. It would be of importance if the data were available as to the optical properties of the phases Dr. Herty finds in the FeO-SiO_2 diagram and in his "synthetic" inclusions. Such information we feel would aid very materially in identifying inclusions, at any rate those of a moderate or large size.

DR. C. H. HERTY: So far as the hydrogen in the steel goes and the silica in solution in the steel goes, I do not believe that you could experimentally determine the amount of dissolved silica in the steel, it would be so small that you could not determine it. But, on the other hand, as I pointed out at the first of the paper, if you did not have SiO_2 in solution in the steel, you would never get silicon in the steel, as you do in an acid furnace when you pick up silicon, because you would be eliminating the SiO_2 as fast as it formed and you would have no back force on the reaction $\text{Si} + 2\text{FeO} \rightleftharpoons \text{SiO}_2 + 2\text{Fe}$ to keep silicon in the bath.

As a matter of fact, in the acid furnace, what probably happens is simply this: when you melt down and come, say, to 0.04 to 0.05 per cent silicon or possibly even lower, your slag is low in dissociated silica on account of the low temperature and the slag composition. Now, as you increase the temperature and increase the silica in the slag, the dissociated silica in the slag increases, which throws more silica into solution in the metal and causes this reaction to reverse, giving you your silicon pick-up. Now, of course, the fact that as you get higher carbons, you deplete the FeO also aids that reaction in coming back, but if you did not have dissolved silica, you could not have silicon in the steel bath.

I might point out that silver chloride is spoken of as being insoluble in water, but actually it is very slightly soluble, and I think it is the same type of thing with silica in steel. You could not measure it directly, you probably would have too many silica inclusions, the silica inclusions in the steel would probably be much higher than the dissolved silica, but the latter is controlling as far as silicon is concerned, and for a given amount of silica, you will have a definite amount of silicon in the metal at a given temperature.

On the hydrogen in the steel, I think you have to get the same answer, and that is simply this, that in order to get the carbon down to a certain point, you have to have a certain amount of iron oxide present. Now, if one heat had twice as much hydrogen as another and the hydrogen affected the oxygen content of the metal, you would have to have twice as much iron oxide in one case in the slag as you would in the other, and we all know that you can come down to about the same carbon content with about the same iron oxide con-

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tent of the slag under almost any condition, that is, for a reasonable range of temperature, and I think in hydrogen you have the same thing as you have anywhere else, you have to suppose the formation of either an iron hydride or water soluble in steel. Water is very soluble in molten slags at very high pressures. There is no reason to suppose that water vapor is not soluble in steel, in very minute quantities, of course, but still soluble enough so the hydrogen in steel would be affected by that amount of water vapor, and the water vapor, iron and hydrogen are in equilibrium with the iron oxide, and the carbon is at the same time, so the hydrogen does not influence the oxygen content such as we have had in these melts.

As to the point that Dr. Hatfield brought out as to the carbon content not being taken into account, evidently Dr. Hatfield misread that diagram we had showing the carbon-silicon type of inclusion diagram. You will notice that those curves swing up to the left, so with 0.06 per cent silicon and low carbons we got a type of inclusion that was high in iron oxide. When you get up to high carbons, 0.25 per cent and up, you get the glassy type.

Dr. W. H. HATFIELD: I was speaking of the oxygen content, not the type of inclusions.

Dr. C. H. HERTY: As you get up to the 0.25 per cent carbon range, you get the glassy type of inclusion, and as you go up in carbon, you come down in

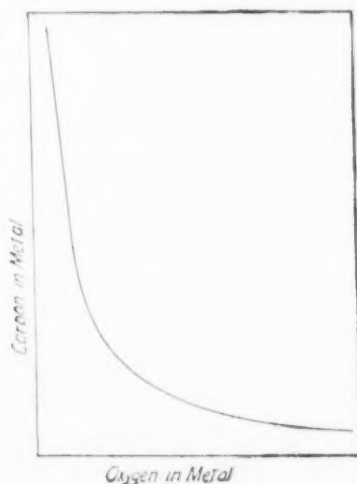


Fig. 1A

oxygen, that is, it is presupposed, of course, that the higher you are in carbon the lower you are in oxygen. As a matter of fact, in the work we have done we could calculate that with a 1 per cent carbon steel we should have about 0.003 per cent oxygen present. When we get down to 0.03 or 0.05 per cent carbon we can have as high as 0.08 oxygen in the steel. That is the range. The general relationship between carbon and oxygen is shown in Fig. 1A. No values have been assigned to the coordinates because the equilibrium constant has not yet been determined.

Now, on the oxygen content of the steels being 0.005 by Dr. Desch's method, I think you will find that is very normal for American practice, or

French or English practice, for that matter, because, after all, equilibrium is independent of country. If you kill a steel with 0.12 per cent silicon the oxygen content would be about 0.009 per cent. If you go up to 0.20 per cent silicon, you would have less oxygen. If you had manganese and silicon, so that you did some fluxing in addition to straight deoxidation, I think you would come down to even lower oxygen contents. Of course, when the oxygen falls as low as 0.004 per cent, the analytical error is fairly large.

As to Mr. Feild's question, I think I covered that more or less when I mentioned silico-manganese as a method of getting a fluid inclusion. Just where the proper ratio of manganese and silicon is to give you manganese oxide and silica in a relationship that will give you fusible inclusions is something that we are working on at the present time and we hope to solve that problem in the next year or so.

In answer to Mr. Cape's question on the petrographic work, we have determined refractive indices on these inclusions. We have checked cristobalite, and from there on up you get a number of different refractive indices, depending on the composition of your inclusion. They go from cristobalite through fayalite and even higher in refractive index.

The other way we have checked the composition is by getting the metallographic properties of the synthetic slags and checking the microstructure of the synthetic slags with the microstructure of the inclusions. For instance, on the two inclusions of the high magnification series, we have checked that structure exactly in the synthetic slags, so right now, on ferrous-silicates alone, we use the microstructure as much as anything else to determine the inclusion composition.

STEEL FAILURES IN AIRCRAFT

By F. T. SISCO

Abstract

This paper discusses failures in steel aircraft parts that have been investigated by the Metallurgical Laboratory at Wright Field. The methods used in the investigation of these failures are described briefly, followed by a classification of the various failures in respect to their origin. Included in this classification is an account, illustrated by typical and representative cases, of: (1) failures due to faulty manufacturing methods and not caused by defective material; (2) failures due to internal defects such as non-metallic inclusions and the like; (3) failures due to seams and other surface defects; (4) miscellaneous failures due to defective steel including large grain size, segregation and banding; (5) failures due to faulty or wrong heat treatment; and (6) failures due to welding.

FOREWORD

IN THE activities of the Materiel Division, U. S. Army Air Corps, the inspection, testing and development of materials and equipment for military aircraft are of first importance. The Materiel Division does not undertake any production work; it does, however, co-operate with the airplane and engine manufacturer in the testing and inspection of raw and finished material as well as in design, to insure complete conformity to War Department specifications. The testing, inspection and research facilities are complete and include everything connected with military aircraft from the small bar of steel or the strand of silk cord, to the complete power plant or the completed airplane.

In order to bring out any inherent or unsuspected weakness in material or design and in order to simulate the most extreme service conditions in a short-time test much of the testing is under more

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A paper presented before the tenth annual convention of the society held in Philadelphia, October 8 to 12, 1928. The author F. T. Sisco, a member of the society, is Chief of the Metallurgical Laboratory, Air Corps, Wright Field, Dayton, Ohio. Manuscript received May 25, 1928.

severe conditions than are ordinarily met with in flight. Consequently failures of material occasionally result. As a typical example of one of these failures the following report of unsatisfactory performance on a 1400 cubic inch displacement 500 horsepower water cooled engine may be cited: "This engine had 9 hours block test at the factory at full throttle, speed range 1900 to 2400 revolutions per minute. Full power calibration runs were made at the power plant laboratory at Wright Field through a speed range of 1800 to 2400 revolutions per minute the duration of which was about three hours. The engine was then used for carburetor tests at the torque stand at various speeds, principally on acceleration tests, for approximately 20 hours. The engine was on magneto tests at the torque stand running between 1800 and 1900 revolutions per minute, mostly at 90 per cent full power when the crankshaft failed. The duration of the magneto tests was 24 hours, 15 minutes. The total time for the engine is 56 hours and 15 minutes. The carburetor test was probably the most strenuous as the throttle was opened wide quickly, from idling speeds to full power and full speed. The nature of this test is such that high torsional stresses are imposed on the crankshaft." Upon examination the crankshaft proved to be defective. It is probable that failure of this shaft would not have occurred under usual flight conditions. On the other hand it is possible that had the crankshaft been free from defects it would have withstood the test satisfactorily.

The work of the metallurgical laboratory at Wright Field is about evenly divided between research on steel and nonferrous alloys, and the examination and diagnosis of metal failures occurring throughout the Air Corps. Failures of material on aircraft in service are comparatively rare; probably 90 per cent of them occur during test. From the large number of failures that have been investigated in the past five years, it has been possible to classify them, to form rather well defined opinions concerning their causes, and to make recommendations to reduce their number.

METHODS USED IN THE ANALYSIS OF FAILURES

The greatest handicap in the investigation of metal failures is that generally all history of the part previous to its acceptance for service by the Air Corps inspector is unknown. It is not often that we know definitely the processes involved in melting, work-

Metallog. 4
Impact 13, 14, 15, 16
Chemical 2
Tension 4, 5
Impact 9, 10, 11, 12
Hardness
Coarse Etch
Impacts 5, 6, 7, 8
Hardness
Coarse Etch
Metallog. 3
Metallog. 2
Coarse Etch
Hardness
Tension 1, 2, 3
Impact 1, 2, 3, 4
Chemical 1
Metallog. 1

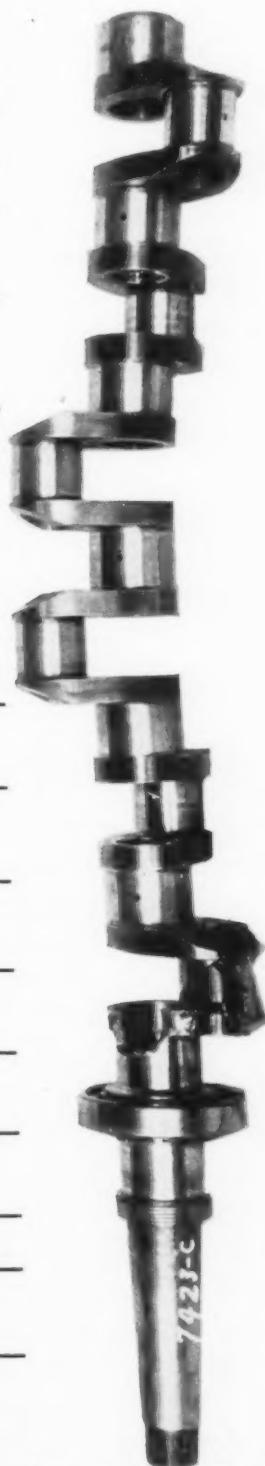


Fig. 1—Photograph of Failed Crankshaft Showing Failure and Location of Test Specimens. (Greatly Reduced.)

ing, and fabrication. Some manufacturers are willing to give a detailed history of the material they are furnishing; most of them however, when they meet the specifications satisfactorily feel that their obligations are ended.

Failures sent in for investigation are accompanied by an unsatisfactory performance report that gives the complete history of the part insofar as this history is known. Nearly always the record of the part is available since its installation in the airplane. At Wright Field the failed part is subjected to a visual examination to determine whether some manufacturing or fabricating fault might be responsible. This includes examination of the fracture to locate its origin with respect to any sharp corners, tool marks or too-small fillets that may act as a notch. Often this visual examination when correlated with a knowledge of the stresses involved will be sufficient to explain the failure. Following this, any or all of the following methods of procedure may be used, chemical analysis, coarse etching, complete physical tests insofar as the shape and size of the piece will permit, metallographic examination, and, if advisable, X-ray radiographic inspection. Fig. 1 shows the location of the specimens in the failed crankshaft described above.

It has been the practice for several years, when engines or airplanes are torn down for a complete overhaul, to make, when possible, a metallurgical examination of parts which have withstood service or test conditions unusually well. By this method considerable information has been obtained that has been of great value in the investigation of like parts that have failed after relatively short life.

Failures of steel parts in aircraft can be divided into two main heads: (1) failures due to faulty design or improper or careless methods of manufacture, and (2) failure due to materials. This classification is necessarily a more-or-less arbitrary one. In the first class we include failures due to overloading or to the part having insufficient strength, or to improper machining; in other words a failure which is not caused by some abnormality or defect in the steel itself or by some abnormality or defect introduced while the material is undergoing a change in structure or composition as in rolling, forging, heat treatment, welding or the like. A failure due to material can usually be traced to some abnormality, a defect either in the steel itself or introduced during the various

processes through which the steel goes before it is ready for the final machining operations.

It is not within the scope of this paper to discuss failures due to design. These are becoming less frequent each year due to the remarkable advance designers have made in stress analysis and their experience in adapting high strength materials to aircraft construction. When it is realized that the average weight of the engine for military aircraft (i. e. the weight exclusive of the cooling system) has decreased from approximately 2.0 pounds per horse power in 1918 to 1.3 pounds per horse power in 1928, or 35 per cent, it is at once apparent that phenomenal strides have been made in engine development. In addition to this reduction in weight the average life of the engine has been doubled; it has increased 100 per cent, with a corresponding decrease in maintenance cost. This almost sensational development in engines for military aircraft—and in the airplane as well—has been due to three factors: (a) improvement in design, (b) improvement in methods for the machining and fabrication of parts and (c) improvement in materials.

Notable among these improvements may be mentioned the successful and wide use of welding; the perfection of precision methods for the manufacture of gears and other highly stressed parts that have been heat treated to a scleroscope hardness of 65 to 75; and the excellence of some of the newer valve steels.

FAILURES DUE TO FAULTY MANUFACTURING METHODS

Before taking up the question of failures due to the material itself, we will pause for a moment to look at a few of the more common manufacturing faults that can be held responsible for failures. It seems almost platitudinous to discuss the notch. The effect of the notch in reducing the resistance of the material to fatigue, to impact, either sudden or repeated, or to vibratory stresses is so well known and so important that it would seem as if this prolific source of failure would have been completely eliminated in modern design. In fact, however, it is quite evident that a few designers do not yet realize the importance of proper filleting because a considerable proportion of material failures in aircraft have their beginning in the sharp corner. A common fault in this respect is not so much the complete absence of the fillet as the

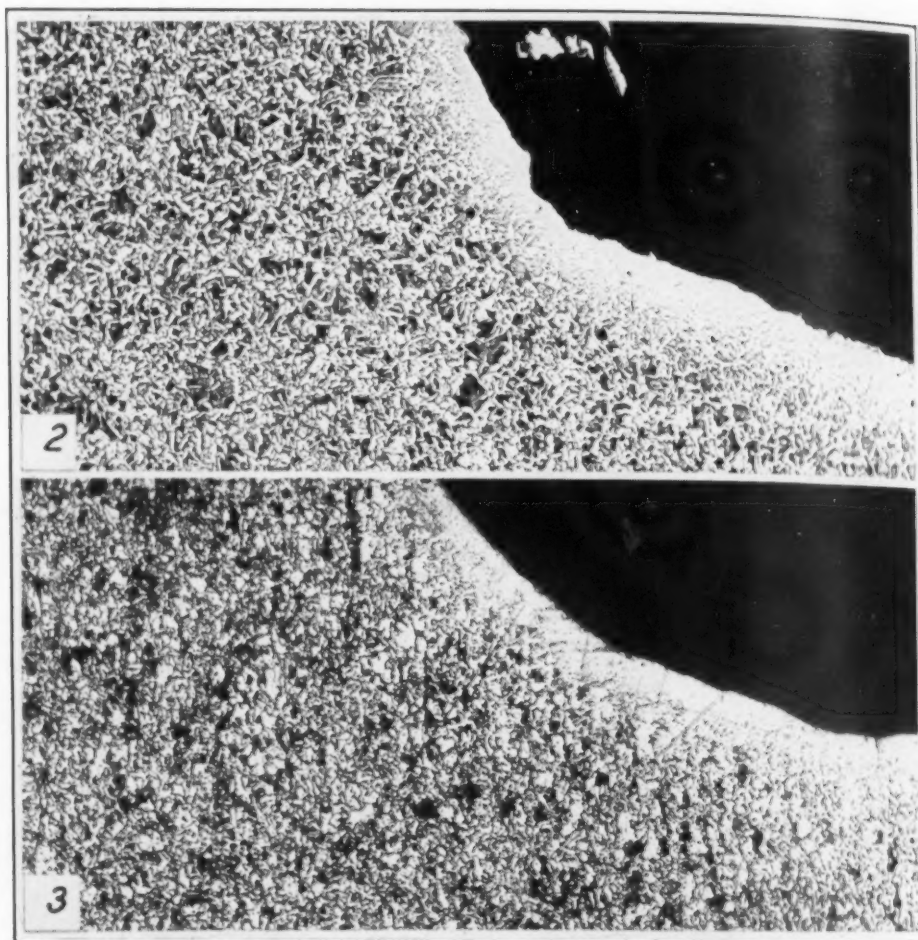


Fig. 2—Photomicrograph of Crankshaft Bolt Showing Irregular Fillet. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 3—Photomicrograph of Crankshaft Bolt Showing Smooth Fillet. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

fillet that is too small. We occasionally encounter drawings on which fillets are $1/64$ inch.

A source of considerable trouble is the fillet that is improperly or carelessly machined so that instead of the fillet performing its proper function it is in effect a notch. This condition was evidenced in the examination of a lot of crankshaft bolts in which several failures developed.

The drawing specified a $1/32$ inch fillet at the head and in the reduced shank portion. A careful inspection showed that all of the bolts were filleted in these two locations yet failures occurred in the fillet of the reduced shank portion.

Metallographic examination indicated that many of the radii

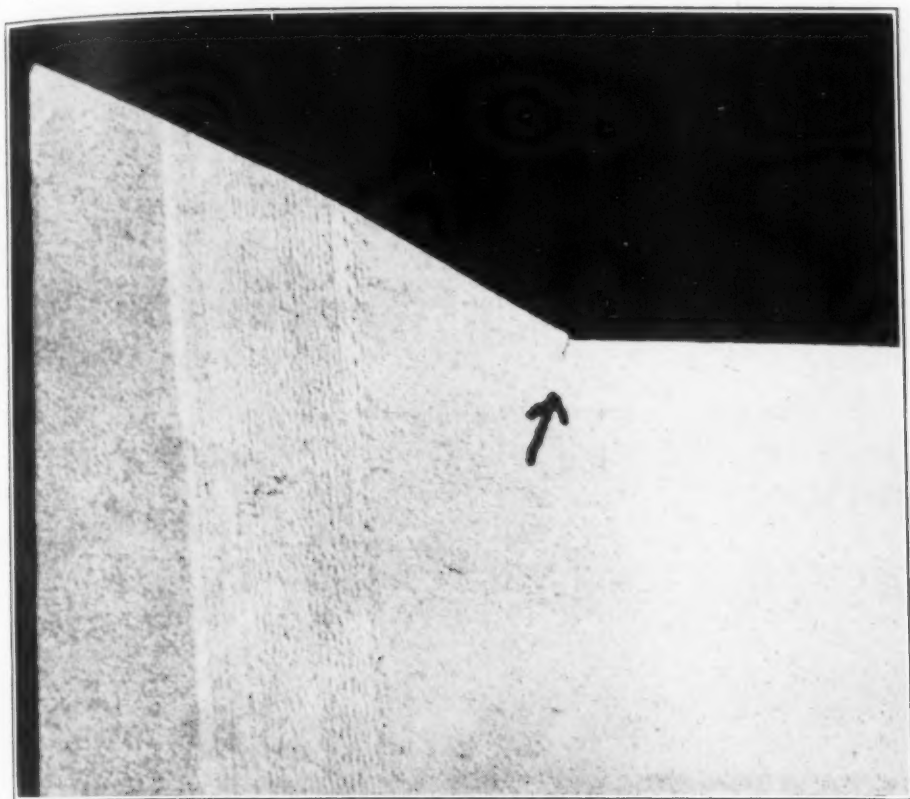


Fig. 4—Section of Propeller Hub Showing Origin of Fatigue Failure. (Slightly Reduced.)

were irregular. The appearance of the fillet in two representative bolts is shown in Figs. 2 and 3. The bolt shown in Fig. 2 failed, the one shown in Fig. 3 was satisfactory. A magnification of 50 to 100 diameters was necessary to detect the irregularity of machining at the fillet. The steel was 3.5 per cent nickel, excellent in quality and heat treated properly and had the fillets been smooth would have undoubtedly withstood the severe vibration satisfactorily.

Fig. 4 shows an interesting example of how a fillet in a soft material may be turned into a notch when in contact with a harder material containing a sharp corner and the whole assembly subjected to severe vibration. Fig. 4 is a cross section of the hub of an adjustable blade duralumin propeller. On all blades of this design there is a liberal fillet between the hub and blade. The steel barrel enclosing the duralumin hub has a fillet at each end which fits into the fillet of the two duralumin blades. Fig. 5 is a sketch of the fil-

let (a) as it appears on the drawing and (b) as it actually had been machined on both ends of this particular steel barrel.

When this propeller assembly was tested the burr on the harder steel barrel forced itself gradually into the softer duralumin thus producing the well defined notch evident in Fig. 4. A crack started at this notch. The result was complete failure of one blade, and the beginning of failure of the other. In Fig. 4, the arrow points to a crack about $\frac{1}{8}$ inch in depth which had started in this blade.

From these illustrations it is evident that to avoid failure in highly stressed parts due to a notch we must watch for other things

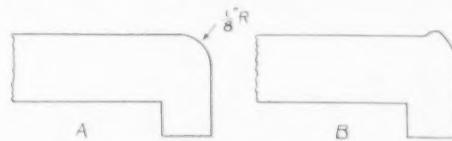


Fig. 5—Sketch of Fillets of Steel Barrel for Adjustable Blade Propeller. (a) Fillet as Shown on Drawing, (b) Fillet as Machined on Barrel. Fillet Actual Size.

in addition to the complete absence of the fillet. Failure may start in a fillet that is too small, it may start in a fillet that has a rough, uneven machined surface, or it may start in a sharp tool mark. In the case of assemblies where two metals, one hard and one soft, are used in conjunction with each other, any projection or sharp edges on the hard metal will produce what is in effect a notch in the softer metal. This is especially true where the assembly is subjected to vibratory stresses.

The importance of careful, accurate and smooth machine work for aircraft parts cannot be too strongly emphasized. In extremely high stressed movable parts the difference between early failure and almost indefinite service depends to a large extent on the care taken in manufacturing processes. The illustrations given in the foregoing paragraphs are only a few, selected to emphasize particularly vital points. Far too many steel failures in aircraft have been caused by a little carelessness in machine work.

GENERAL DISCUSSION OF FAILURES DUE TO FAULTY MATERIAL

In discussing failures due to faulty material, two divisions naturally suggest themselves. In the first place the steel may be inherently bad; it may contain slag, blowholes, or other internal defects; or it may contain seams or other surface imperfections.

In the second place it may have been improperly processed at some stage of its manufacture and before the finish machining operations and final assembly in the completed airplane or power plant. It may be improperly rolled or forged, or it may be improperly heat treated. We can conveniently include welding in this second classification.

In spite of the advances in the steel maker's art, especially in the manufacture of alloy steels, some dirty steel is still being made, and of this dirty steel considerable finds its way into aircraft construction. In spite of the improvement and perfection of the various processes to which the steel ingot is subjected before it appears as the finished part, there is occasional evidence that faulty rolling mill and forging practice still exists. There is still careless heat treatment being done,—it is not frequent,—but it becomes evident occasionally in a failure, especially of a carburized part.

Fig. 6 is plain evidence of faulty mill practice. This part is a valve spring washer for a radial air-cooled engine. After coarse etching a long narrow rectangular area was noted. This is plainly shown in Fig. 6. Microscopic examination showed that this area was a low carbon steel strip. Fig. 7 at a magnification of 100 diameters shows the junction of the heat treated nickel-chromium steel washer and the low carbon steel strip. It is quite evident from Fig. 7 that carbon has migrated from the steel of the washer into the low carbon strip. This leads to the supposition that the defect probably originated in the ingot.

Some steel makers using big-end-up molds for alloy steels insert a hook made of mild steel into the molten metal in the top of the ingot. When the ingot solidifies this hook facilitates stripping. Cases have been known where in rolling the ingot it was not cropped sufficiently and a small piece of this hook remained in the ingot. This was probably what happened in this particular washer; it was machined from the top end of the top billet. A condition such as this might well be dangerous. Had an oil or water pump drive gear or other vital, highly stressed part been machined from such steel as this a premature and perhaps disastrous failure might have resulted.

Happily such examples of faulty mill practice are comparatively rare. Most mills are careful about their cropping but the detection of just one instance of such flagrant carelessness is evidence

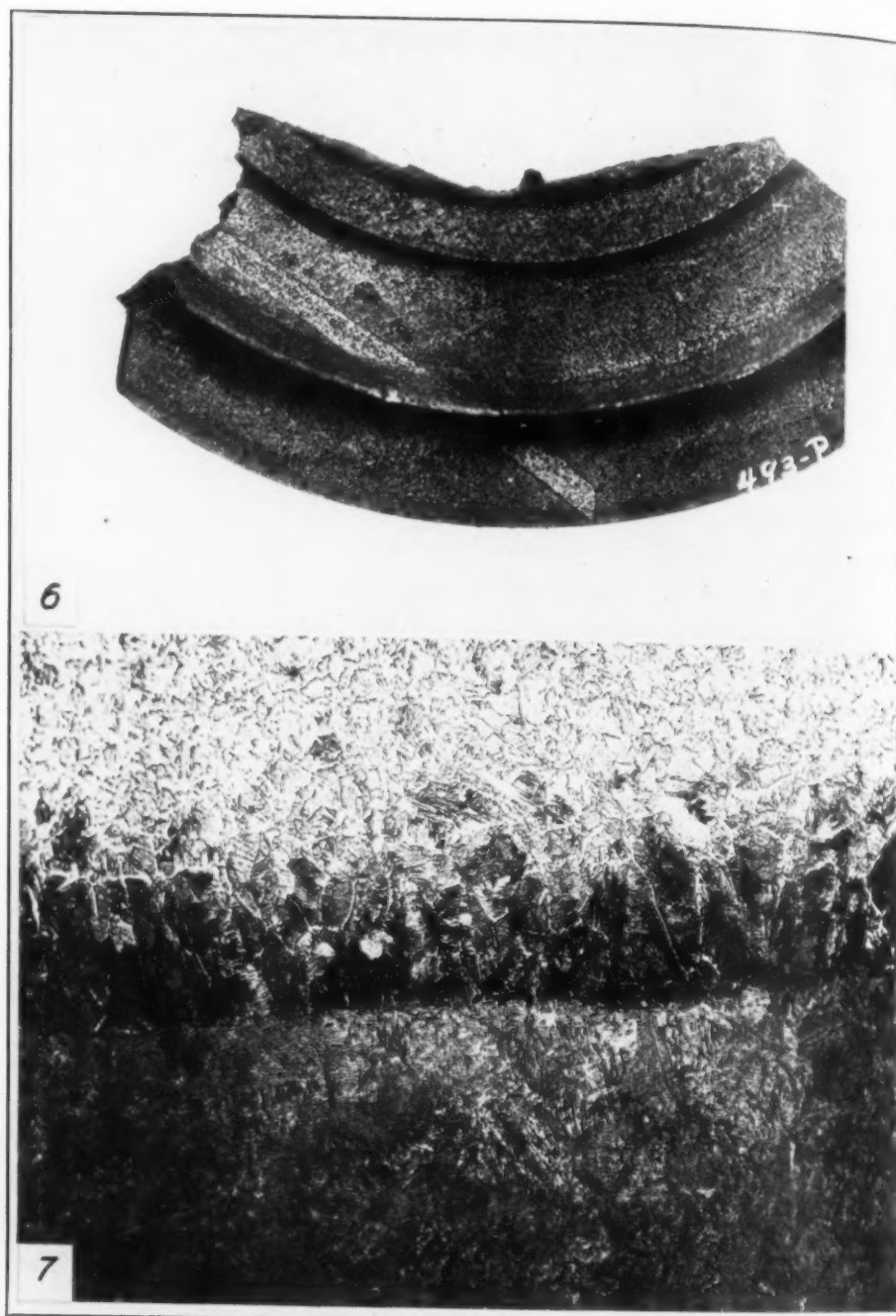


Fig. 6—Photomacrograph of Low Carbon Area in Valve Spring Washer—Coarse Etched. $\times 2\frac{1}{2}$. Fig. 7—Photomicrograph of Junction of Steel of Washer and Steel of Low Carbon Area Shown in Fig. 6 Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

that such defects may be encountered even in the highest grade alloy steels.

FAILURES DUE TO SLAG AND DIRTY STEEL

Slag inclusions, blow holes and internal seams in alloy and other high strength steels, assume an importance in aircraft parts far greater than usually accorded to these internal defects. Most failures of aircraft engine parts take place by fatigue and are the result of a variable or alternating load. It is well known¹ that "if any portion of the metal, however small, where the concentration of stress occurs, is stressed beyond its fatigue limit, it will fracture after a time under alternating stress and thus cause a minute crack. The concentration of stress at the bottom of this crack will be greater than before and the process will repeat itself, and in this way the crack will spread and ultimately cause fatigue fracture of the whole."

It has been established without doubt that a concentration of stress occurs at sharp corners, keyways, screws threads, toolmarks, gear teeth and the like and leads to early fatigue failure. There has in the past been some question as to whether non-metallic inclusions accelerate failure by fatigue. The results brought out by Lucas² in his study of the path of fatigue failure and by Styri in his discussion of Lucas' paper, show conclusively that fatigue cracks frequently have their origin in clusters of slag particles. In this connection Moore³ says: "A very few tests have been run on specimens of 'dirty' steel.... The test data for these few tests are erratic. The effect on resistance to fatigue of any inclusions in steel seems to be very largely a matter of chance, depending on whether the inclusion is located at a point of high stress or at a point of low stress. Some specimens of dirty steel gave as high fatigue test results as did specimens of clean steel, while other specimens of dirty steel gave low results. A study of the fatigue strength of dirty steel might be expected to yield results of great importance...."

In the work done at Wright Field on failures a mass of evidence has been accumulated which shows without doubt that non-

¹Hourwich and Foster—Air Service Engine Handbook, Materiel Division, Wright Field, 1925

²TRANSACTIONS, American Society for Steel Treating, April, 1927, page 531.

³Moore and Jasper, University of Illinois, Vol. XXI, No. 39, May 26, 1924, Bulletin No. 142, page 65.

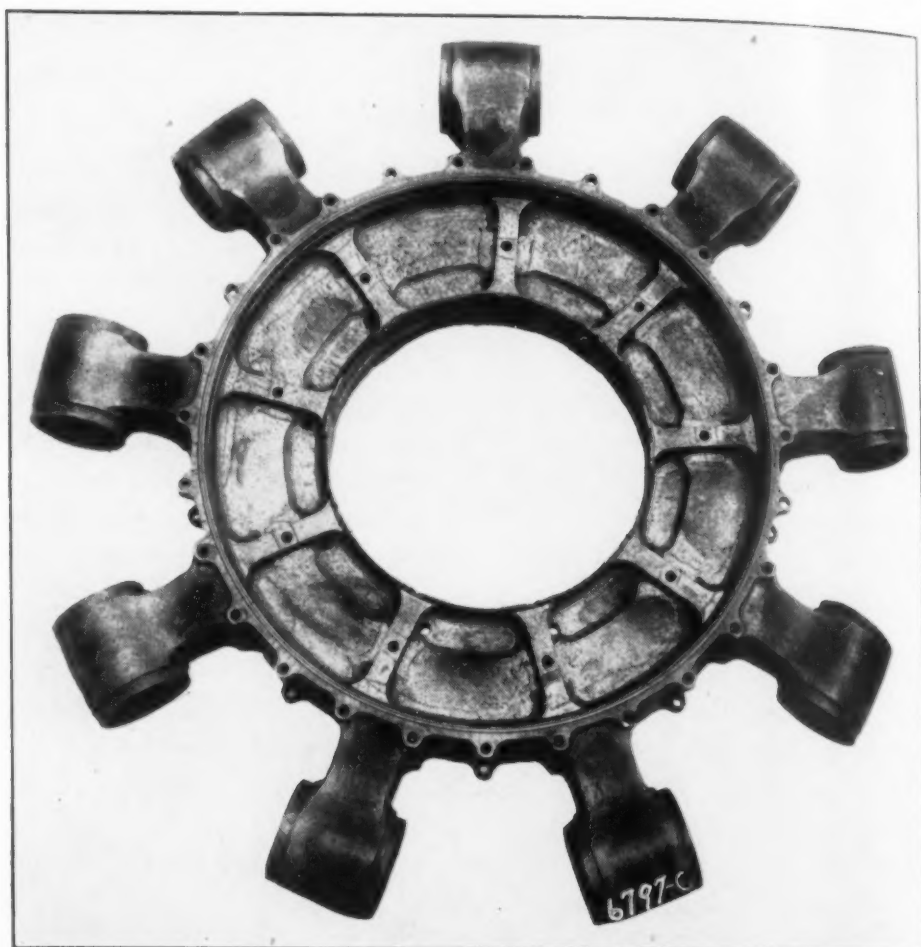


Fig. 8—Wobble Plate for Barrel Type Engine (Greatly Reduced). Coarse Etched.

metallic inclusions are frequently responsible for, or at least accelerate fatigue failure. This evidence has been so strong that engine builders are now paying an almost exaggerated attention to the cleanliness and quality of the steel.

An excellent example of early fatigue failure in exceedingly dirty steel is shown in Figs. 8, 9 and 10. Fig. 8 is a wobble plate from a 425 horsepower experimental barrel type engine. This plate is subjected to relatively high alternating stresses, but stresses which are probably much below the fatigue limit for this material, which is a 0.35 per cent carbon, chromium-vanadium (6135) steel heat treated to 260 Brinell hardness.

A test on this engine was discontinued due to a bearing failure after 10 hours at part throttle and 11 hours at full throttle.

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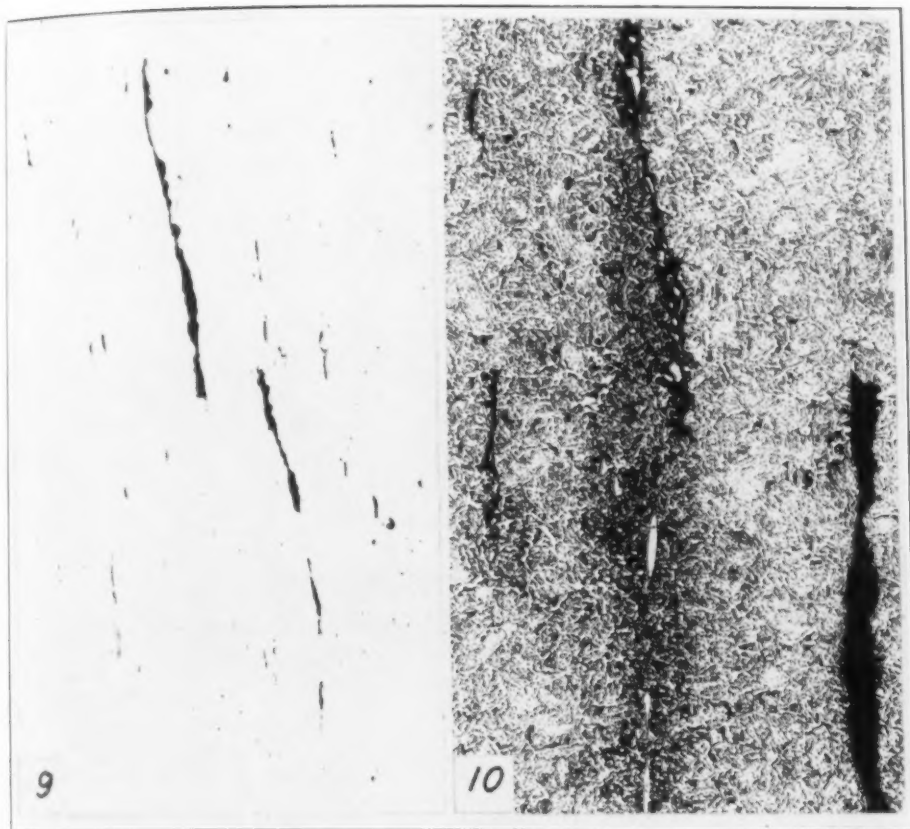


Fig. 9—Photograph of Slag Particles in Specimen Through One Rib of Wobble Plate. Specimen Unetched. $\times 35$. Fig. 10—Photomicrograph of Path of Fatigue Crack Through Inclusions in Wobble Plate. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

Examination of the wobble plate when the engine was torn down disclosed a number of small cracks in the fillets of the central ribs. The part was then coarse etched with the result that more than 100 cracks were detected. Every fillet contained one or more. The larger of these cracks are enclosed in pencil marks and a few of them can be seen on Fig. 8. Figs. 9 and 10 are representative of a large number of metallographic specimens taken from this part. Fig. 9 shows the character of the slag and Fig. 10 the path of a fatigue crack progressing through particles of slag.

The value of the evidence obtained by the examination of this part was especially great in that the piece was available for examination just before failure had actually occurred. When the failure of a highly stressed moveable part does take place, it (and sometimes the engine) is usually completely wrecked. In any case



Fig. 11—Photograph of Failed Cylinder for Water Cooled Engine Showing Failures. (Greatly Reduced.)

its condition is such to make the cause of failure somewhat uncertain.

Fig. 11 shows a chromium-molybdenum steel cylinder that failed in a 600-horsepower engine after 41 hours endurance testing at 2500 revolutions per minute. The structure adjacent to the one end of the crack in the cylinder shown in Fig. 11 is shown in Fig. 12 and the other end of the crack in Fig. 13. The stresses on the walls of this cylinder are such that it is quite logical to assume that there was a concentration of stress in a colony of slag inclusions.

The examples given above could be duplicated almost indefinitely. Turning now from the experimental metallurgical evidence to the more practical, and perhaps just as conclusive, evidence the following experience may be cited. Several engine build-

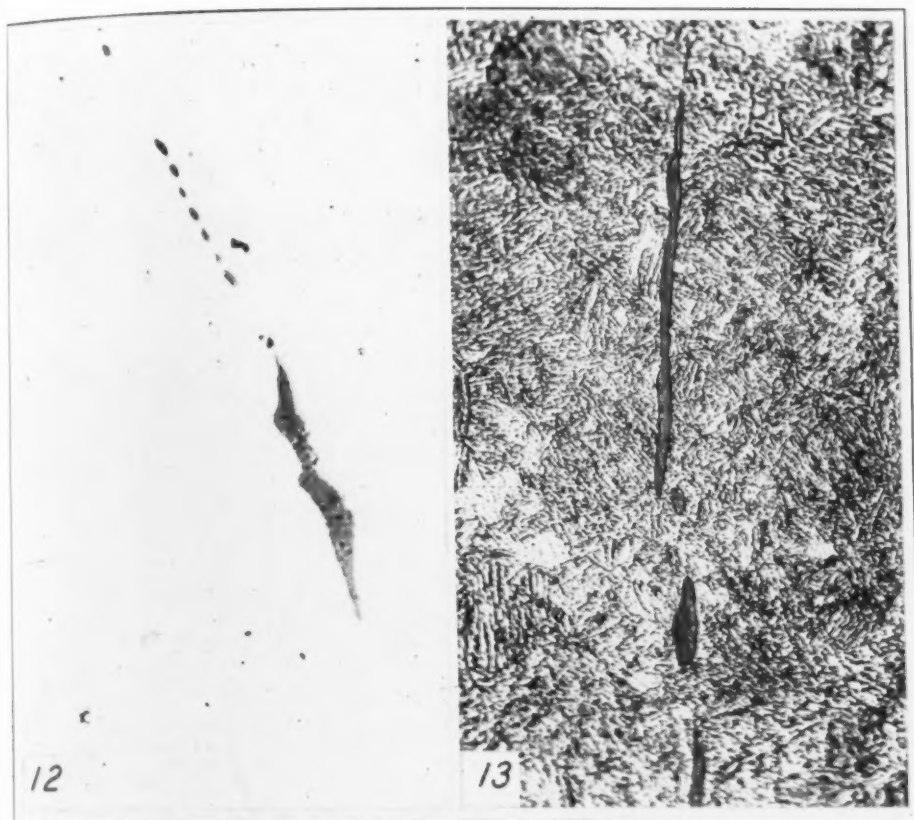


Fig. 12—Photomicrograph of Slag Particles at One End of Crack in Cylinder. Specimen Not Etched. $\times 100$. Fig. 13—Photomicrograph of Slag Particles at Other End of Crack. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

ers were troubled by an excessive number of valve spring failures. These springs were made out of a chromium-vanadium spring wire by experienced spring makers. The stresses on a valve spring can be estimated with a fair degree of accuracy. In one particular engine the stress in the valve spring varied from approximately 30,000 to 50,000 pounds per square inch. So far as is known the torsional endurance limit for spring-tempered chromium-vanadium steel has not been determined, but past experience has set 30,000 to 50,000 pounds per square inch as a safe working stress for these springs.

When an excessive number of failures developed, the springs were increased in size so that the stress range was decreased to approximately 20,000 to 35,000 pounds per square inch. Even with greater margin of safety the number of failures did not diminish noticeably.

An investigation disclosed that the steel in the springs was not of high quality; the detection of slag inclusions and other defects led to the conclusion that the failures were due to defective steel. A search was made for a source of supply of high grade steel and finally through the co-operation of a piano manufacturer, high

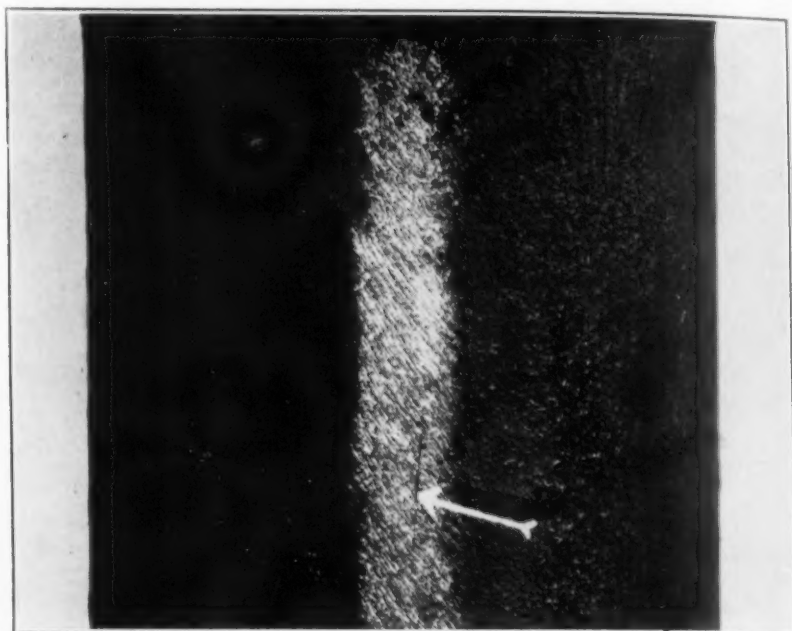


Fig. 14—Photomicrograph of Oil and Water Pump Drive Gear Showing Surface Seams. Specimen Coarse Etched. $\times 2\frac{1}{2}$.

carbon piano wire was used for valve springs. With the introduction of these springs the stress range was increased to 35,000 to 50,000 pounds per square inch and the springs placed in service. Failures of these springs are almost unknown.

Sufficient airplane engine valve springs which have failed in service and in test have been examined metallographically so that coupled with the experience of engine builders on this vital part, it is safe to assume that slag inclusions and other small internal defects will cause premature failure, and that a valve spring made of clean steel will give satisfactory service.

FAILURES DUE TO SEAMS AND OTHER SURFACE DEFECTS

When we attempt to fasten the blame for material failures on seams and other surface imperfections we find ourselves on still more debatable ground than the question of blaming the fail-

ure on non-metallic inclusions and like internal defects. It is true no doubt, that in many cases a seam will act as a notch producing a concentration of stress at this point and resulting in a premature failure.

Figs. 14 and 15 show an oil and water pump drive gear from a V-type air-cooled engine which had been subjected to a 10-hour

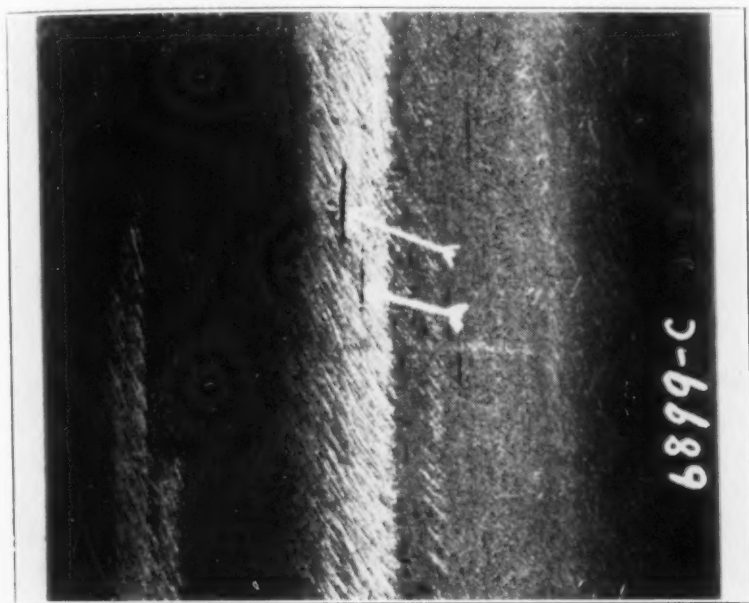


Fig. 15—Photomicrograph of Shaft of Oil and Water Pump Drive Gear—Showing Defects. Specimen Coarse Etched. $\times 2\frac{1}{2}$.

full throttle test at the factory and 12 hours at part throttle at Wright Field. At the expiration of this test the throttle was opened wide and after one hour the gear failed. Coarse etching revealed a large number of seams of which some are evident in Figs. 14 and 15. Several seams were evident in the fillet between the hub and the flange. There is little doubt but that a concentration of stress occurred in one of these seams and caused failure.

One of the most important and yet most debatable problems this laboratory has faced is the question of crankshaft failures; that is whether or not defects in the steel can be blamed for causing or contributing to the failure. An airplane engine crankshaft is subjected to high stresses; and stresses so complicated and with so many factors entering, that they cannot be analyzed with any degree of certainty. There is no doubt but that there is a concentration of stress in the fillets; and crankshaft design has, as a

result, had as one object a stiffening of the shaft to reduce this concentration of stress. The shaft is at times subjected to abnormally high stress due to resonant vibration at certain speeds and especially to preignition, a condition more apt to occur in supercharged engines.

The question of whether or not defective material enters into a crankshaft failure is almost wholly of the harmful effect of hair-line seams. Forging practice on crankshafts is entirely satisfactory; as is heat treatment and the resulting properties and structure. The only defect encountered in most crankshafts is the small surface imperfection known as the hair line seam. This seam is frequently little more than a scratch; being from $\frac{1}{8}$ to $\frac{1}{2}$ inch in length and often of no measurable width or depth. They often occur in colonies; and few crankshafts have been examined that have been wholly free from them. In the majority of cases crankshafts fail in the first or second throw from the propeller end. Fig. 1 shows such a typical failure. The stress on the throws is a combination of bending and torsion and is greatest at this point. Figs. 16, 17, 18 and 19 show sections from various crankshafts that have failed prematurely. Fig. 16 shows the journal of the throw that was third from the failure. There was a colony of small seams in the fillet, one of them large enough to be seen in the photograph. Fig. 17 is the journal of the throw next to the failure in the same shaft. In this journal a crack has started in the fillet and has progressed radially around the journal, turning at the end so that it is apparently headed for the oil hole. Fig. 18 shows the journal of the throw next to the failure of another crankshaft. It will be noted that the crack has started radially in both directions from the oil hole. Fig. 19 shows the appearance of a section of the crankshaft from a small experimental two cylinder engine. The sponginess and the large seams in the steel are readily apparent.

Practically all crankshaft failures are due to fatigue; most of them have their origin either in a fillet or in an oil hole. There is little doubt but that the direct cause of failure is a highly concentrated stress at the point where failure starts. That this failure is due to overloading is probable. There is no direct evidence that hairline cracks, small seams or other defects cause the failure but there is considerable basis for the assumption that they contribute materially.

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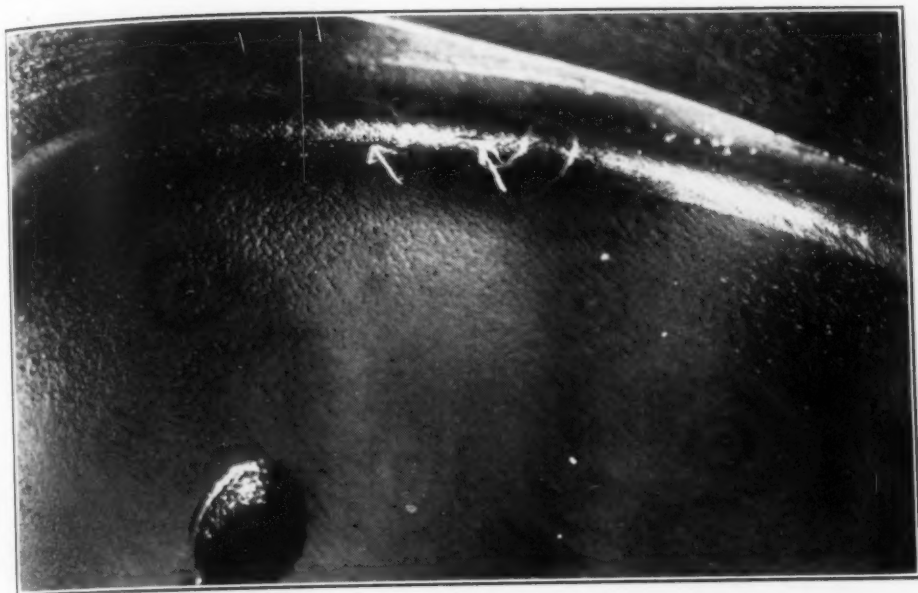


Fig. 16—Photograph of Fillet of Crankshaft Journal Showing Small Seams. (About Actual Size.)

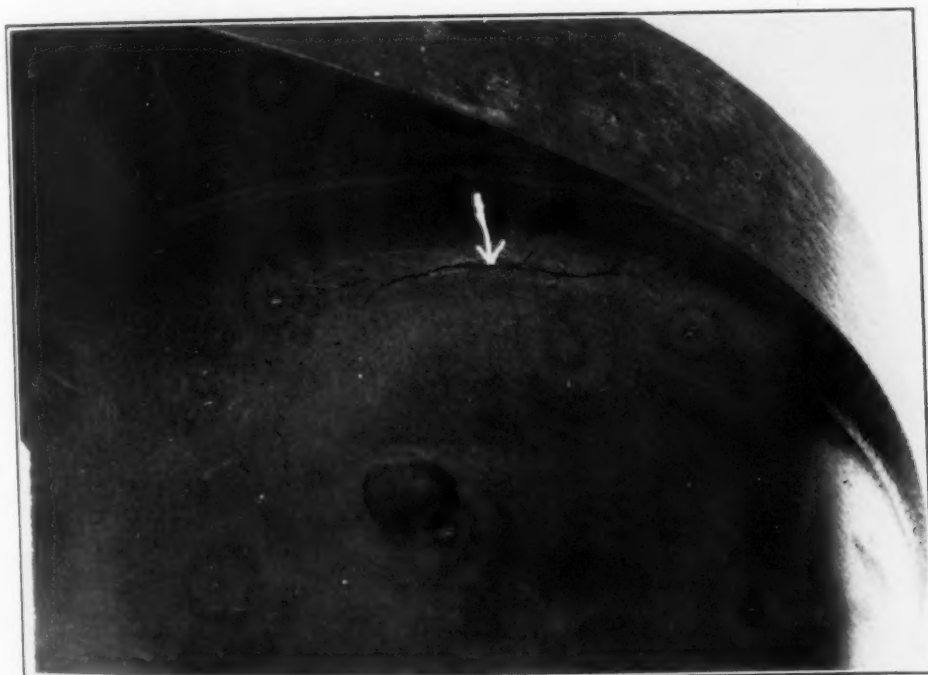


Fig. 17—Photograph Showing Fillet of Crankshaft Journal Showing Crack. (About Actual Size.)

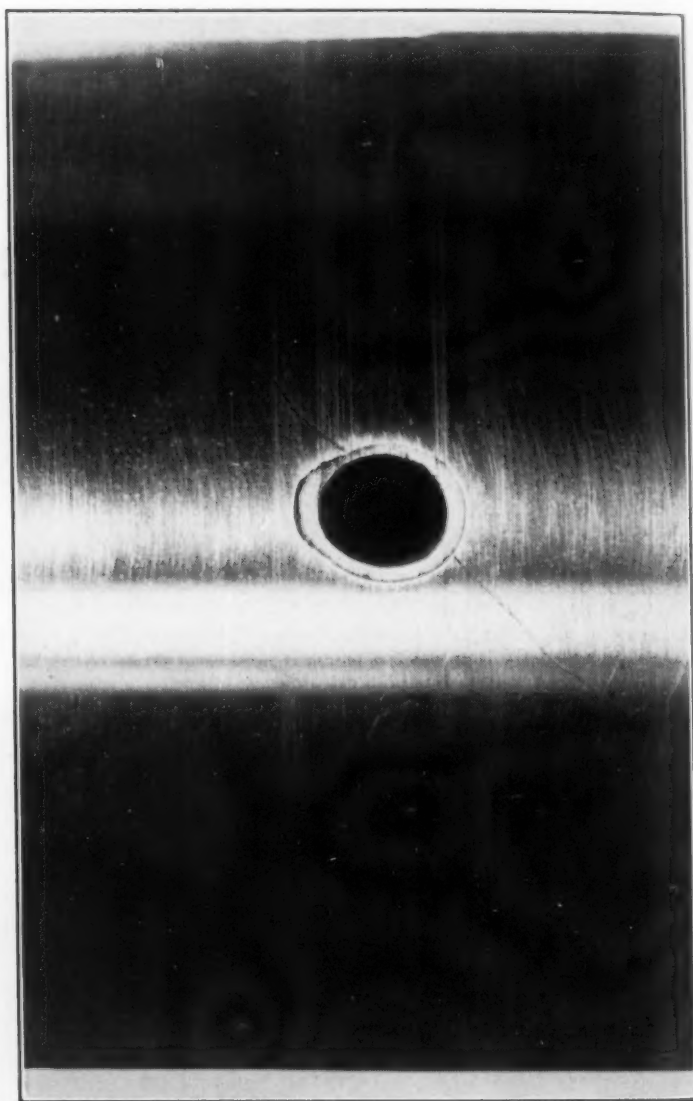


Fig. 18—Photograph of Crankshaft Journal Showing Crack at Oil hole. (About Actual Size.)

Most likely the large seams and cracks shown in Figs. 16, 17, 18 and 19 have been opened up by the high stress at the time the shaft failed, but the presence of numerous hairlines in the rest of the shaft and in locations adjacent to where these cracks were found makes it logical to assume that the large seam or the sizable crack had its origin in the hairline seam. Hundreds of crankshafts that probably contained many of these hairline seams have given excellent service. On the other hand we have no evi-



Fig. 19—Photograph Showing Defective Steel in Small Engine Crankshaft.
(Reduced.)

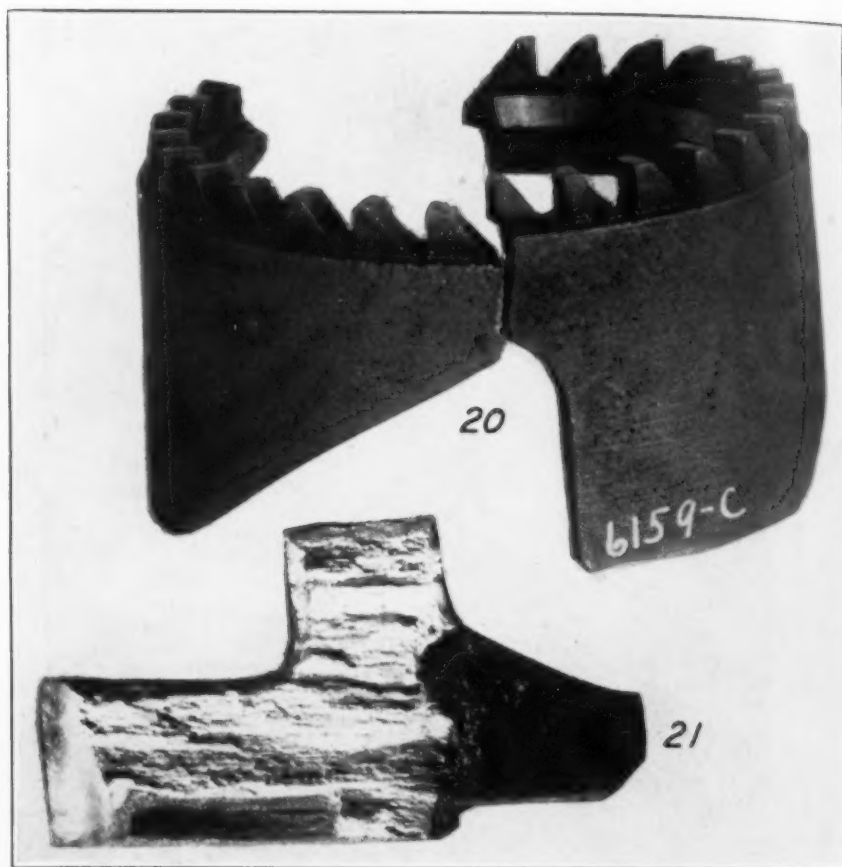


Fig. 20—Photograph of Hand Starter Release Showing Location of Fracture, $\times 1\frac{1}{2}$. Fig. 21—Photograph Showing Appearance of Fracture of Hand Starter Release, $\times 6$.

dence that these shafts have been subjected to any abnormally high stresses. It is true that a crankshaft might fail due to a concentration of high stresses and still be free from hairline seams, but the preponderance of evidence is that for a given stress the shaft that contains large numbers of hairline seams and other small surface defects will fail where the shaft wholly free from these defects will not fail.

MISCELLANEOUS FAILURES DUE TO DEFECTIVE STEEL

A few failures in aircraft parts can be traced to the presence of defects in the steel other than non-metallic inclusions, and surface seams and cracks. Fig. 20 shows a failure due to woody structure. This part is a release on a hand operated engine starter and

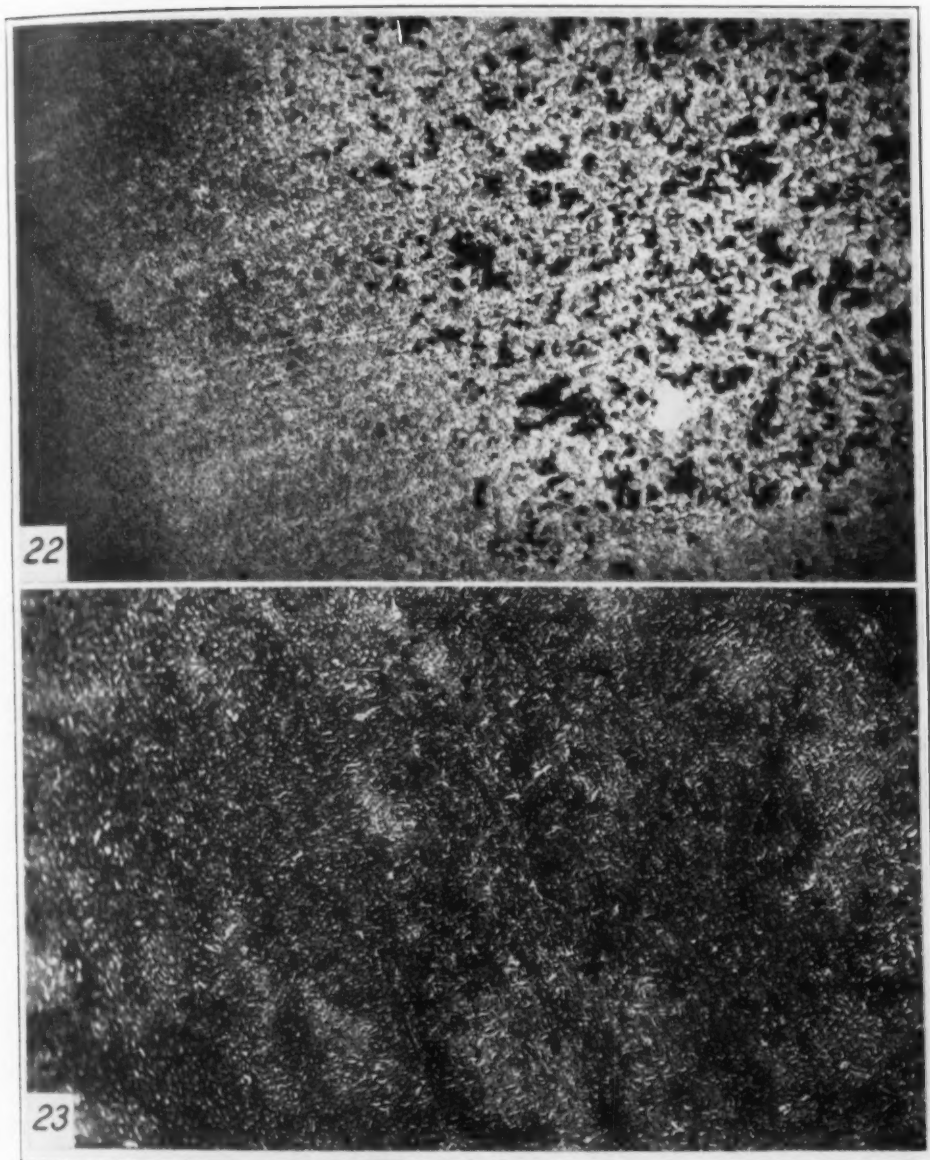


Fig. 22—Photomicrograph of Structure of Hand Starter Release Intersecting Fracture. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 23—Same as Fig. 22, Specimen Parallel to Fracture. Specimen Etched in Alcoholic Nitric Acid. $\times 40$.

failed as shown in Fig. 20. The woody appearance of the fracture is evident in Fig. 21 and the heterogeneous structure in Figs. 22 and 23. This type of failure was much more common several years ago than at the present. Along with flake, woody fracture has almost completely disappeared; very few cases have been encountered in the last few years.

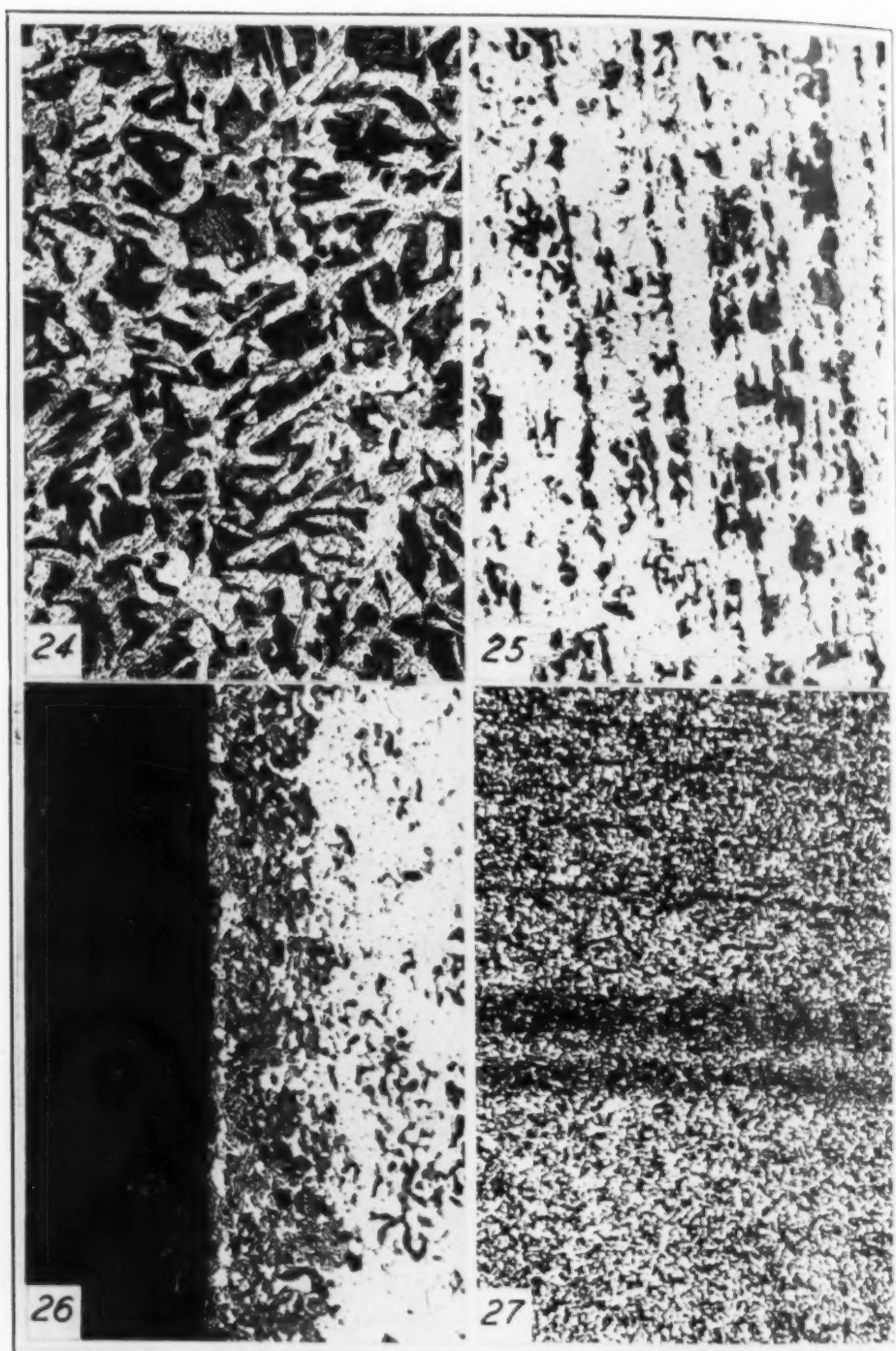


Fig. 24—Photomicrograph of Failed Wheel Flange Showing Large Grain Size. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 25—Photomicrograph of Strut Fitting. Average Structure. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 26—Photomicrograph of Strut Fitting. Segregated Area Near Edge. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 27—Photomicrograph of Failed Bomb Rack Fitting, Showing Banded Structure. Specimen Etched in Alcoholic Nitric Acid. $\times 50$.

Occasionally a failure can be blamed on abnormal grain structure, but on the other hand the abnormal structure can often be associated with non-metallic inclusions or is due to heat treatment, or sometimes to the absence of heat treatment. It is only necessary to cite one or two instances where failure can be traced to the structure. Fig. 24 shows the average structure of a 0.45 per cent carbon steel forged flange from an airplane wheel. The flange failed prematurely. This part is subjected to unimportant tension stresses and also to rather severe impact stress in case of a rough landing. When the part is forged and heat treated according to specifications no difficulty is encountered. In the case of the piece whose structure is shown in Fig. 24, the part was evidently not heat treated at all. The grain was coarse and the material too soft. The result was early failure.

Figs. 25 and 26 show a case of an unusual failure due to abnormal structure. The part was a strut fitting made from low carbon sheet stock. Fig. 25 shows the average structure of the material and Fig. 26 a representative area near the edge where there was considerable carbon segregation. Several of these segregated areas were noted in the specimen examined. Evidently the sheet stock from which this fitting was made was rolled from a segregated billet. Failure was caused by vibration and started in one of these segregated areas.

As yet it has been impossible for this laboratory to associate banded structure and ghost lines definitely with the failure of steel parts, but from the large number of failures examined in which banded structure was the only abnormality present and in which it is known that the stresses probably were not excessive, it seems safe to assume that there is some relation between this abnormality and premature failure. Fig. 27 shows a typical banded structure in a failed bomb rack fitting where no appreciable amount of slag could be detected in or near the bands, even at relatively high magnification. As the phosphorus in this particular steel was 0.021 per cent it is hardly possible to blame the failure or the banded structure on this impurity.

We might say just a few words about the use of the wrong steel for highly stressed parts. Even though unintentional, it seems inexcusable, but the fact remains that it happens occasionally. As example of this we might cite the failure of a magneto shaft for

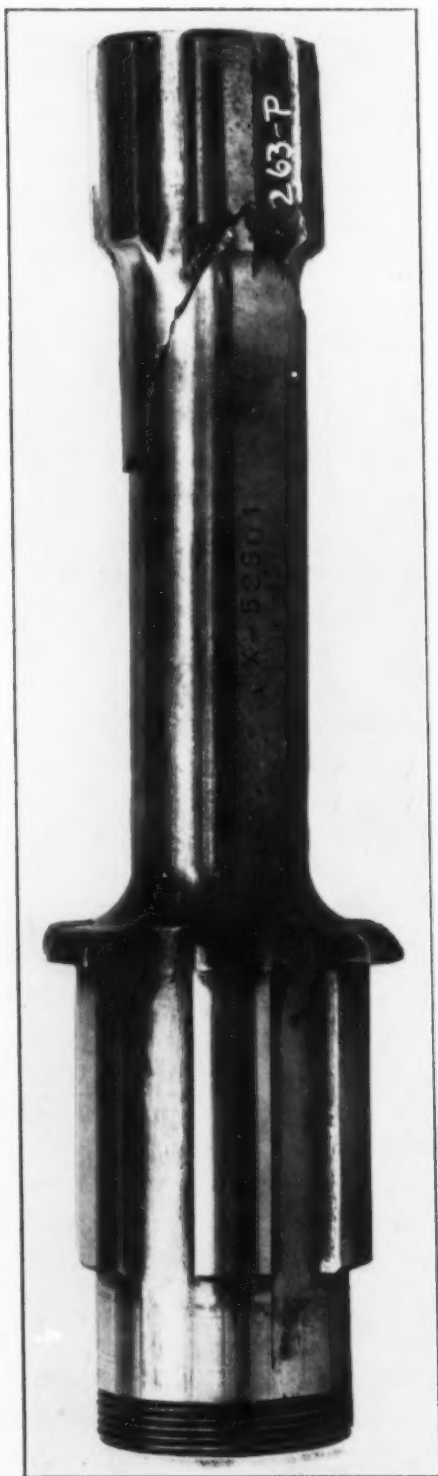


Fig. 28—Photograph of Supercharger Drive Shaft Showing Failure. (Slightly Reduced.)

which a Bessemer steel, free-cutting screw stock containing 0.095 per cent sulphur and 0.087 per cent phosphorus was used. Another case was the use of a silico-manganese steel containing 2.0 per cent silicon for a supercharger drive shaft for which a 0.50 per cent carbon, chromium-nickel steel heat treated to 60 scleroscope hardness was specified.

FAILURES DUE TO HEAT TREATMENT

The effect of erratic or wrong heat treatment on the life of highly stressed steel parts is too well known to demand anything but a passing attention. Failures have been caused and probably always will be caused occasionally by subjecting the part to the wrong treatment; for example, the part is too hard and brittle and fractures by a relatively light impact, or the part is too soft and wears rapidly.

Due partially to the care used by most manufacturers in heat treatment and partially to the realization of the importance of generous fillets wherever possible, failures due to hardening cracks are comparatively rare.

Fig. 28 shows a supercharger drive shaft that failed during the acceptance tests of the engine. It was machined from a 0.65 per cent carbon, 1.55 per

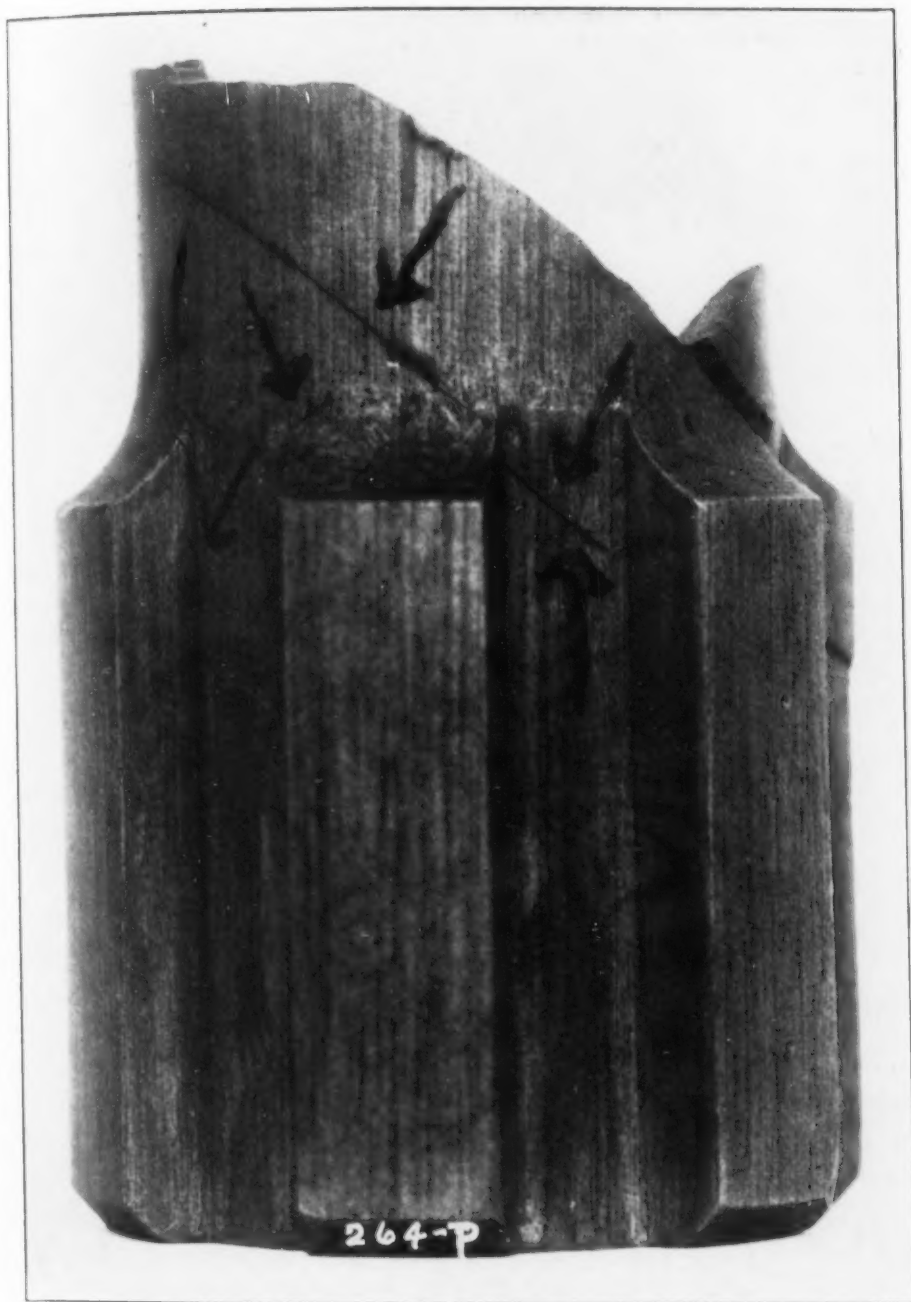


Fig. 29—Photograph of Spline End of Failed Supercharger Drive Shaft Showing Hardening Cracks, $\times 1\frac{1}{2}$.

cent nickel, 1.10 per cent chromium steel of fairly good quality. The carbon content was 0.10 per cent above the upper limit specified for 3250 steel. Fig. 29 shows the spline end after coarse etching. There

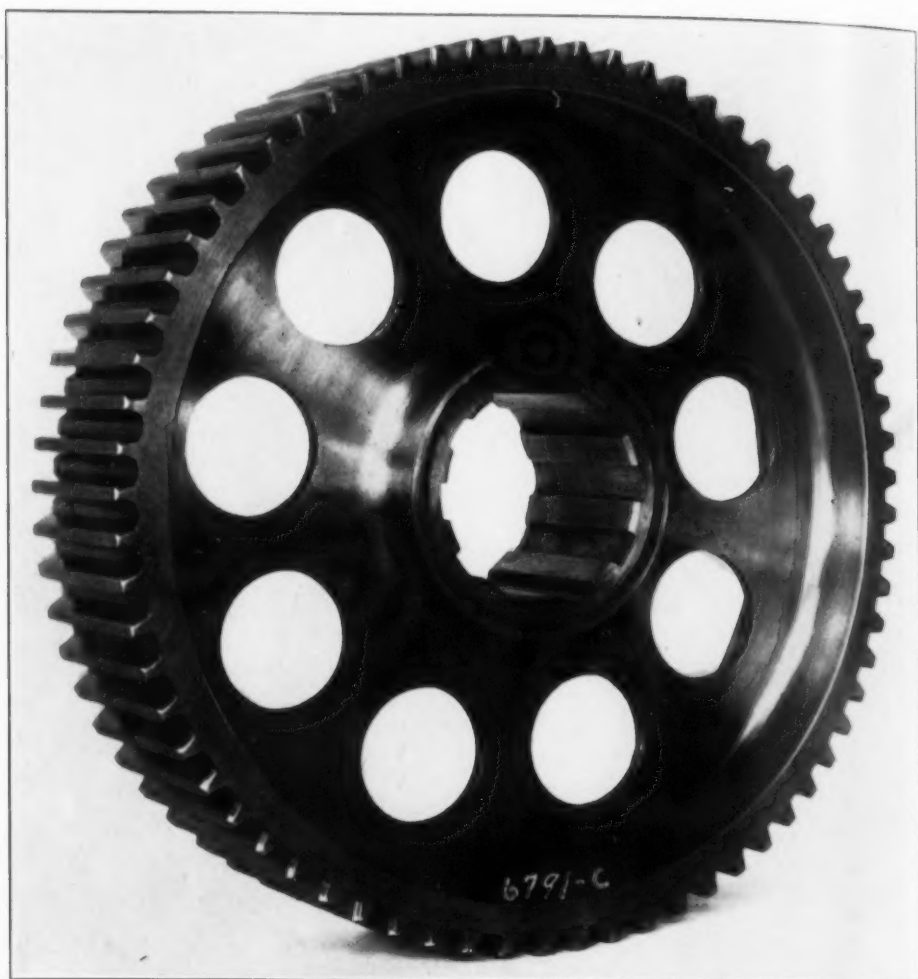


Fig. 30—Photograph of Reduction Gear for Geared Propeller. (Greatly Reduced.)

were hardening cracks at the base of each spline. Many of these hardening cracks extend beyond the end of the spline and change to a radial direction, evident in Fig. 29. The failure was due to torsion and no doubt originated in a hardening crack. Three factors probably entered into the formation of these hardening cracks: (1) the heat treatment necessary to attain a scleroscope hardness of 60; (2) the unusually high carbon content for a chromium-nickel steel; and (3) the exceedingly small fillets at the base of each spline.

As representative of failures due to improper carburization one instance will be given. Fig. 30 shows a reduction gear for a geared propeller. This gear was made from a low carbon 5.0

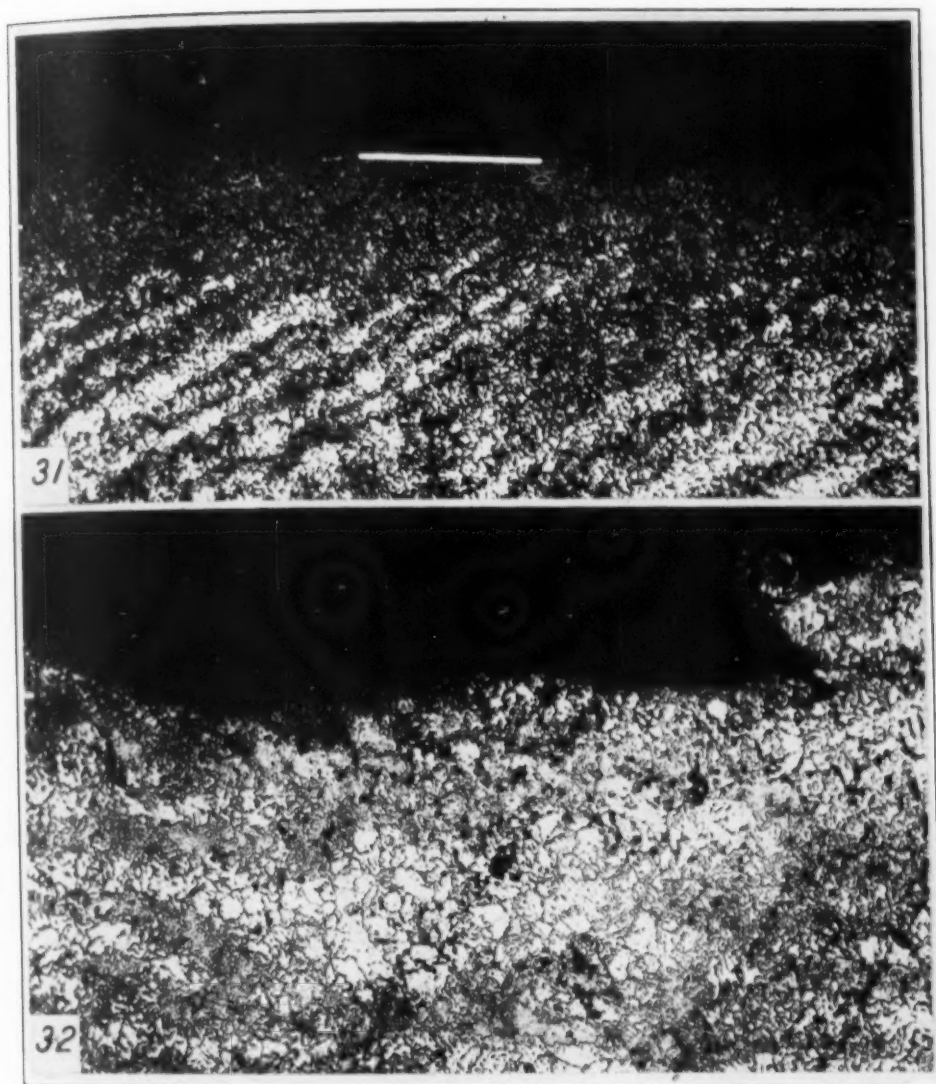


Fig. 31—Photomicrograph of Tooth of Reduction Gear. Not Pitted—Showing Depth and Structure of Case. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 32—Photomicrograph of Tooth of Reduction Gear. Badly Pitted, Showing Absence of Case. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

per cent nickel steel supposedly carburized to a depth so that the case would be 0.020 to 0.030 inch after grinding. Fig. 31 shows the structure of a tooth that was not pitted and Fig. 32 the structure of one in which severe pitting occurred. The case in Fig. 31 is less than 0.01 inch, where 0.020 to 0.030 was specified; and in Fig. 32 it is quite evident that there is no case at all. Most failures of carburized parts occur for one of two reasons: (1) the piece was carburized unevenly or machined unevenly or perhaps

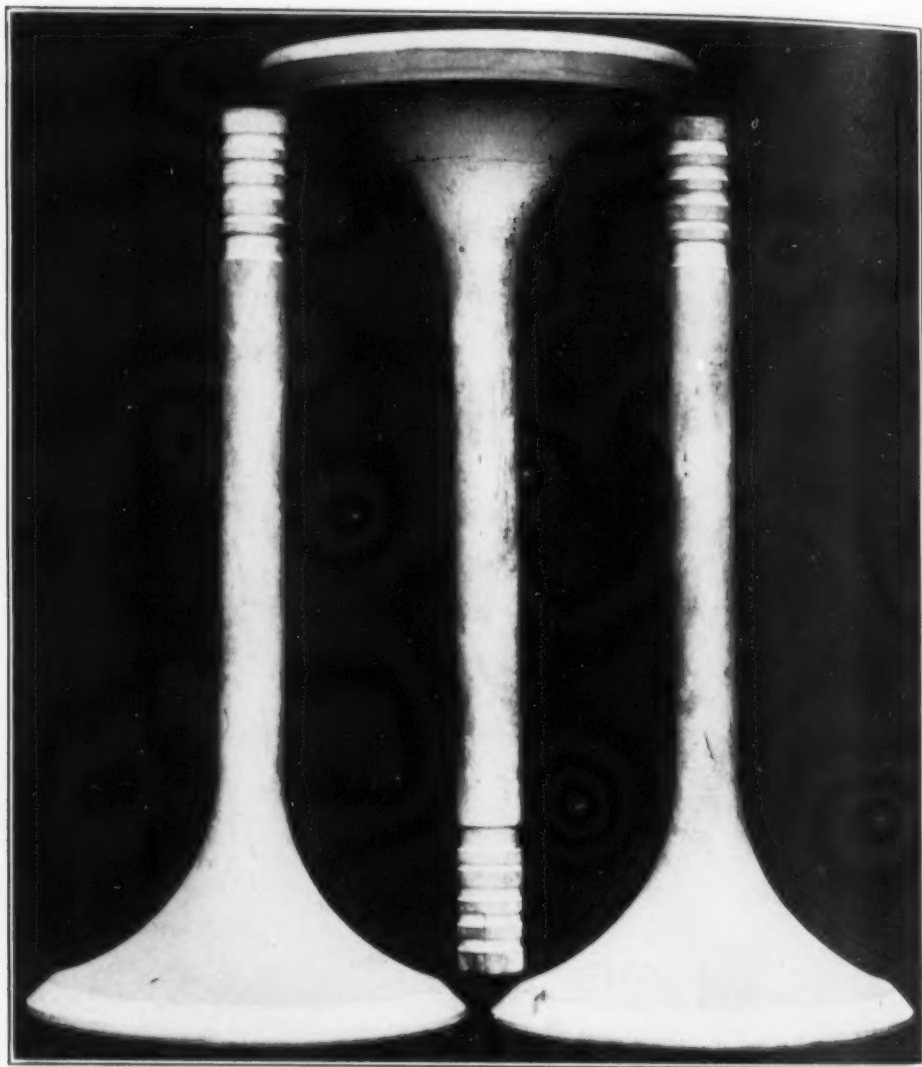


Fig. 33—Photograph of Valves, Coarse Etched and Zinc Plated, Showing Surface Cracks, Slightly Reduced.

distorted so that practically all of the case was removed from portions of the piece in the grinding operation or (2) the case was uneven and heterogeneous in structure or the core was improperly refined and was coarse grained and brittle. These cases of failure are all due to erratic or improper heat treatment.

An unusual case of failure due to a combination of slightly unsound steel and too severe heat treatment was encountered in a rather important valve failure. These valves were made from a 0.58 per cent carbon, 0.71 per cent chromium, and 1.82 per cent

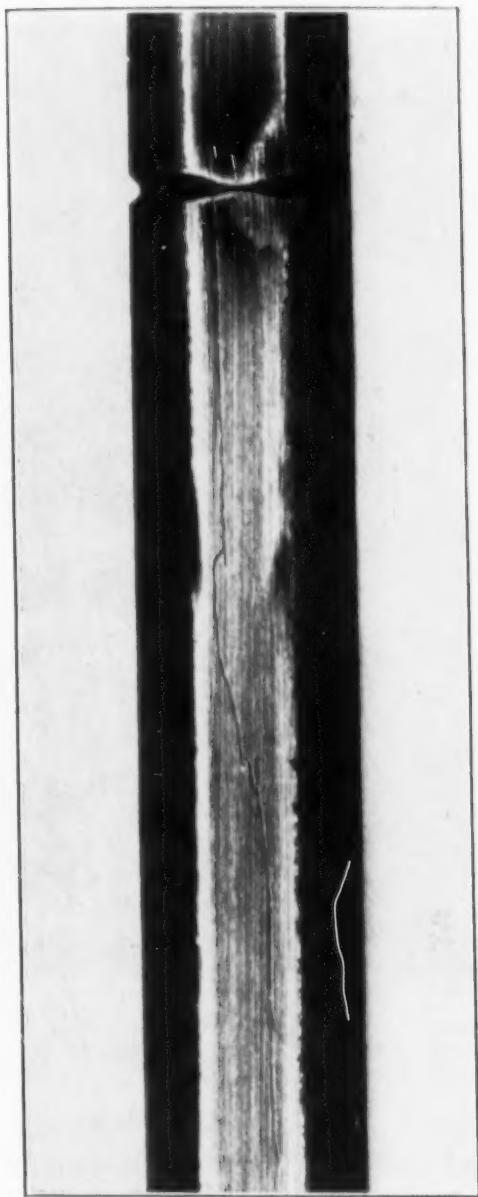


Fig. 34—Photograph of Bar Stock for Valves—Coarse Etched, Showing Crack. (About Actual Size.)

tungsten steel heat treated drastically in order to attain a hardness of 80 scleroscope or greater. Failure of two valves in an engine test lead to the examination of the remaining valves of the lot. Fig. 33 shows three of this lot after coarse etching and zinc plating. The valve at the left has two cracks about $\frac{1}{2}$ inch long in the

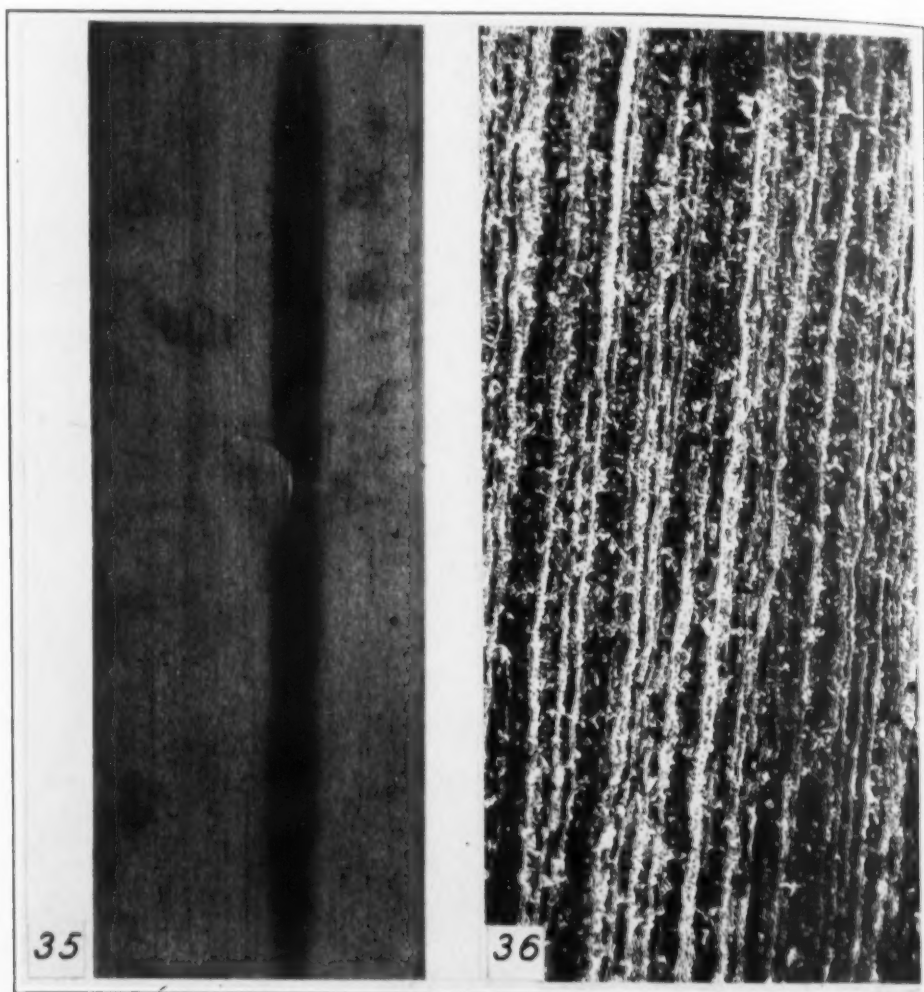


Fig. 35—Photomicrograph of Bar Stock for Valves—Polished and Etched in Alcoholic Nitric Acid—Showing Unsound Area. $\times 3$. Fig. 36—Photomicrograph of Structure of Unsound Area Shown in Fig. 35. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

stem, one at the middle, the other about an inch from the head. In the center valve circular cracks will be noticed in the head, and in the valve at the right, a deep crack $\frac{1}{4}$ inch long and one inch from the head is seen. These were unused valves.

The metallographic structure of the valves was somewhat heterogeneous but nothing could be detected that gave any clue to the origin of the defect. A bar of the steel from which the valves were made was found to be apparently sound as were the rough forgings. After considerable investigation it was found that the cracks were due to the drastic heat treatment and were

nothing more than areas of intense strain which opened up as a crack upon deep etching in hot acid.

The source of the cracks was determined by experiments on the bar stock from which the valves were made. A number of specimens from this stock were quenched in oil from 1625 degrees Fahr. Some of these bars were tempered immediately at 300, 400, 500 and 600 degrees Fahr. Others were held 24 and 72 hours be-

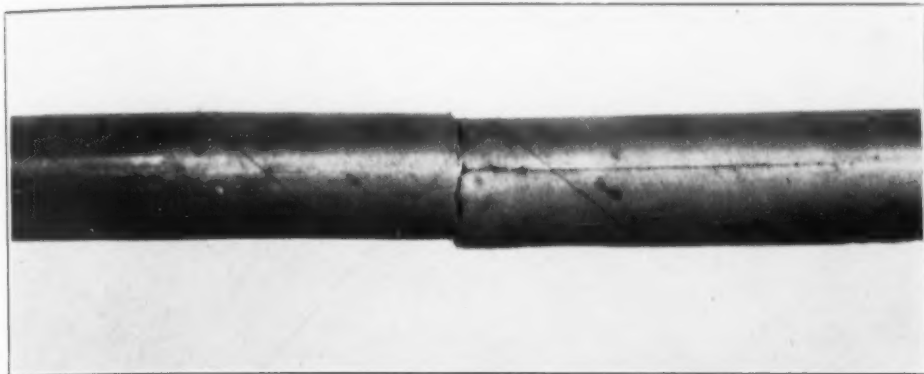


Fig. 37—Photograph of Failed Shaft from Dual Stick Control. (Greatly Reduced.)

fore tempering at the same temperatures. All of the specimens were then polished and coarse etched. The bars hardened and tempered immediately at 400, 500 and 600 degrees Fahr. were apparently free from cracks. Some of the bars tempered immediately at 300 degrees Fahr. contained small cracks, others contained none. Practically all of the bars held for 24 or 72 hours before tempering contained cracks, some of them very large.

When these bars were lightly etched in hot dilute hydrochloric acid small surface seams were detected. Upon further etching these small seams developed into large cracks, one of which is shown in Fig. 34. This gave further plausibility to the assumption that the cracks were caused by an area in which severe hardening strains were present, possibly even a large number of minute hardening cracks. When etched in hot acid these areas developed into a crack the size of which depended to some extent upon the severity of the etching.

Deep etching of the bar stock from which these specimens were taken failed to reveal any cracks, seams or other defective areas in which these hardening cracks could start. But upon polishing a section of the bar stock carefully and etching lightly in 2

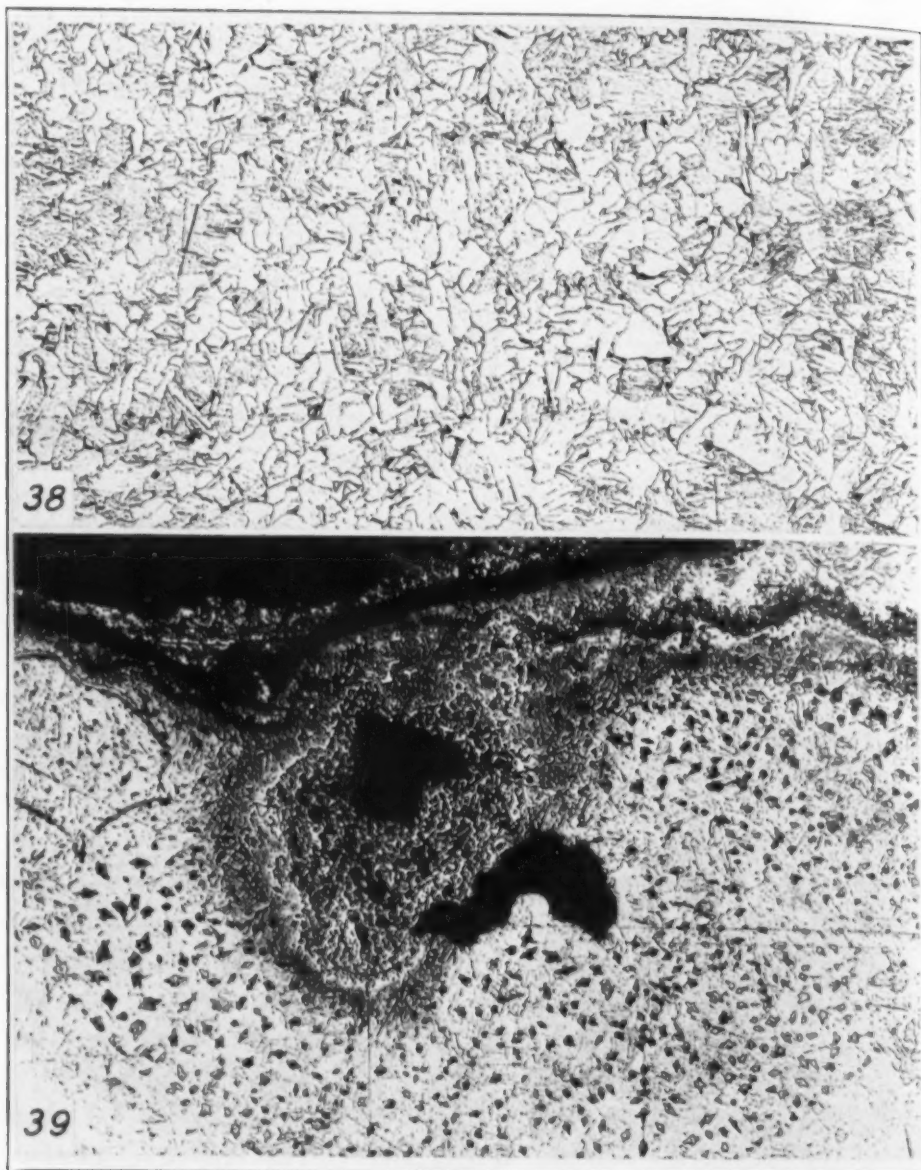



Fig. 38—Photomicrograph of Average Structure in Sound Weld. Specimen Etched in Alcoholic Nitric Acid. $\times 100$. Fig. 39—Photomicrograph of Structure in Unsound Area of Weld. Specimen Etched in Alcoholic Nitric Acid. $\times 100$.

per cent alcoholic nitric acid for two minutes the area shown in Fig. 35 became evident. Further investigation proved that these areas were numerous in the bar stock. Metallographic examination made evident the banded structure shown in Fig. 36 which was characteristic of the dark area of Fig. 35. The rest of the bar was not banded to any appreciable extent.



From this it is quite evident that the cracks in the finished valves which were revealed only by deep etching were due to the drastic heat treating necessary to attain a minimum hardness of 80 scleroscope and had their origin in an unsound area in the original bar stock. This investigation brings out two important things both of which have already received attention but which should be investigated further. These are: (1) the danger of drastic hardening without immediately tempering to relieve hardening strains and (2) whether or not deep etching actually exaggerates a small defect and magnifies it out of all proportion to its original importance and whether or not these small defects would be magnified in importance, and thus made a basis for rejection.

FAILURES DUE TO WELDING

The art of welding especially of thin tubular products has progressed to the point where it is practically 100 per cent perfect. The perfection attained is so high that at present the welded fuselage is standard in airplane construction. Other applications of welding in airplane and engine construction are too numerous and too well known to demand further consideration. Failures due to improper welding in the construction of the airplane are exceedingly rare, when they do occur they are almost always due to the welder and seldom if ever have their origin in the materials or the process.

One or two failures in a mill-welded tube have occurred that warrant attention. For some construction the mill-welded tube may be satisfactory; for the airplane however, it has been necessary to supplant this product by the seamless tube. Fig. 37 shows the failure of a shaft from a dual stick controlled airplane. This shaft is used to connect the control stick in the front cockpit with the one in the rear pit. The failure was transverse, but was accompanied by a longitudinal crack starting at the failure and running along the weld for a distance of 8 inches (Fig. 37).

The specification for the mill-welded tube states that "a test specimen shall be crushed endwise until its outside diameter is increased in one zone by 25 per cent or until one complete fold is formed. The specimen must meet this test without cracking." The specification further states that one test specimen shall be taken from each 100 feet of tubing.

Three test specimens taken from this failed tube passed the crushing test satisfactorily and one failed. A complete metallographic examination was then made, specimens being taken at intervals in the sound portion of the tube. Fig. 38 shows the average structure of the weld. The structure is homogeneous, sound, and the grain size not excessively large. Fig. 39 shows an area of the weld which was decidedly unsound. An unsound area such as this can be detected only by accident, either by metallographic examination or by a crushing test.

From the examination of this failure it is evident that unsound areas may be present in a mill-welded tube and may cause premature failure. It is evident that it is almost impossible to detect a faulty tube unless the inspector is extremely lucky in picking a test specimen that contains an unsound area. Since mill-welded tubes have been replaced by seamless tubes failures have naturally ceased.

ACKNOWLEDGMENT

The writer wishes to acknowledge the assistance and co-operation of N. N. Tilley, mechanical engineer, engine design branch, Wright Field, in the preparation of this material.

DISCUSSION

Written Discussion: By T. W. Downes, metallurgist, U. S. Naval Aircraft Factory, Philadelphia.

In addition to being of considerable interest to the layman or casual reader because of the specific instances of failures of aircraft parts which are described and analyzed therein, the author's paper constitutes a valuable contribution to this department of the literature in that practically all of the known and suspected causes of failure of such parts are discussed. This should be of great help to many whose work includes that of investigating failures of stressed parts, particularly those who are just entering or who are comparatively new in the field of physical metallurgy.

Most failures of aircraft engine parts, as pointed out by the author, occur by fatigue due to repeated stresses. According to the condition of the part as regards internal stresses, the magnitude of the externally applied repeated stresses necessary to cause failure in a given time may be large or small. Quenching for hardening induces internal stresses. These are relieved progressively by tempering as the tempering temperature is raised, presumably never being completely eliminated, so that there always remain internal stresses in hardened and tempered parts. The point is that it is easily possible and probably often true that early failure of parts is promoted by the existence of

internal stresses which act in the same direction as the externally applied stresses, thereby serving to amplify the latter. This condition may easily be completely overlooked particularly in the case of parts the hardness of which may indicate great strength but in which are present great stresses having their inception in hardening and persisting because of insufficient tempering. In this connection is mentioned the fallacy of developing by means of too drastic heat treatment a hardness greater than that ever intended for the particular grade of steel, a practice frequently resorted to but which cannot be too strongly condemned.

In connection with the subject of internal stresses and their probable effect in hastening failure of parts subjected to vibration or repeated reversal of stresses a rather interesting investigation in which the writer participated is considered worthy of mention. The part under test was a streamline wire which was known to have "fluttered" rather severely before failure and one object of the test work was to determine whether or not internal stresses existed in the wire, and if so, to determine the extent thereof. Several specimens about ten inches in length were cut from the wire and thoroughly coated with paraffine. One edge of each specimen was then cleaned of paraffine by scraping and the specimens were immersed in a suitable solution of nitric acid until slightly less than one-half of each wire had been removed. By wet grinding after pickling the irregular edge left by pickling was made smooth and the total portion of the wire removed by both pickling and grinding was made exactly one-half. During grinding care was taken to avoid heating the wire. It was found that during removal of the one-half of the wire the remaining half had acquired a pronounced bow, the thin edge being convex. The cold rolling of the wire apparently induced stresses in the streamline portion which were manifested by curvature when the wire was split along its minor axis. By regarding the half-wire as a beam subjected to uniform bending moment calculations indicated longitudinal compressive stresses in the extreme fiber of the thin edge of the wire of over 50,000 pounds per square inch. Further investigation showed that suitable tempering largely removed the stresses. An estimate of the exact extent to which the internal stresses figured in the failure of the wire is not hazarded but that these contributed to the cause of failure is not doubted.

Written Discussion: By Lieutenant Commander William Nelson (CC), U. S. Navy.

I have read Mr. Sisco's article on "Steel Failures in Aircraft," and regard it as being very comprehensive in that it gives definite indications of failures in steel, from which all ought to be able to take a lesson. It is unfortunate that Mr. Sisco could not go further into the failures of steels used in the structure of aircraft. The future undoubtedly holds a considerable amount in store regarding the use of steel in aircraft structures. Wood and aluminum alloys now form the major part of structures designed and undertaken, but steel, particularly the high tensile steels, are becoming of more interest to the aircraft designer.

It is my opinion that the major developments of steels in the aircraft industry have exceeded the economic use of these materials. In other words,

there are sufficient types of steels to meet practically every problem that comes up. This would tend to indicate that the next steps advisable for those responsible for this great advantage, are to reduce the cost and improve those materials that are now available. Minor improvements and minor reductions in cost will undoubtedly carry a great weight. Naturally, the improvements referred to particularly, are those which will reduce steel failures.

There are a number of minor items which warrant attention, but which on the whole are relatively neglected, all of which lead eventually to failures in the material built into units, such as aircraft. The principal items in this classification are:

- (1) Unsatisfactory means of identification;
- (2) Lack of proper inspection;
- (3) Improper treatment.

Careful attention to the identification of material after it reaches the user is an absolute essential, in order to avoid the incorrect use of steels. It frequently happens that the markings on steels become erased by the elements or by other means, and that the part marked does not have the identification affixed in such a way that when a piece is cut all parts remain properly marked. The identification frequently goes with the part to be used, leaving the remaining piece unmarked in the storehouse. This leads to dangers. Some satisfactory means, and preferably a standard one, of permanently marking steel is of utmost importance.

The introduction of the metallurgist and metallographer into the inspection of steels is considered to be an absolute essential to the reduction of failures. The ordinary physical tests placed on specimens do not always indicate full characteristics, so that failures will be obviated. The metallurgist, through his superior attention to details and his knowledge of the results to be obtained, can improve the inspection of steel material. The inspection of steels under a glass in addition to the machine tests, is particularly important in the alloy materials and demands the fullest consideration.

The inspection of steels after they have been built into the finished unit is relatively neglected. There are a considerable number of items which happen during the operation of an airplane, which warrant a running inspection to avoid unnecessary hazards. A considerable number of standard procedures have been in use by operators of aircraft, and the extension of those procedures seems advisable, if we are to obtain all the information that comes through the observation of units in service.

Mr. Sisco has given some of the results that come through improper heat treatment. It is believed that they are only a part of the improper treatments to which materials are subjected. Treatments include all manner of working of the material from the raw stock to the completed article; these are many and varied, and quite involved in some cases. Full knowledge of what the trade is doing, will generally lead the manufacturer in the right direction. Pending full knowledge of the item under consideration, it is believed that a conservative attitude towards the treatment of the material should be followed. It is rather difficult to indicate definite examples of the

matter concerned, but in many cases, on a study of this subject, it has been found that material is worked too strenuously, resulting in undesirable factors in the completed unit.

The number of hazards which exist in aircraft are great without introducing the hazards connected with faulty material or workmanship. It therefore behooves all those concerned to utilize every effort in reducing the chances of material failures. The stresses placed on aircraft structures and engines are such that fatigue must be given a position of major importance. It may be in some cases that fatigue is the final cause of failure, whereas the original cause is due to something else. Complete analysis of failures in aircraft materials are of utmost importance to obtain the object lesson that undoubtedly exists.

There does not seem to be any doubt but what the aircraft industry will, through its own development, cause improvements in the materials involved to the benefit of many other industries.

Written Discussion: By Horace C. Knerr, consulting metallurgical engineer, Germantown, Philadelphia.

Mr. Sisco's paper is a valuable contribution not only to the aircraft industry but to the entire metal-working field because of its great educational value in the prevention of failures of metal parts in service.

It is unusual to meet with such a frank and comprehensive discussion of failures and their cause, with illustrative examples, even in the papers of the American Society for Steel Treating, as ordinarily the latter have been restricted to some particular type of failure or phase of the subject.

It is through a prompt and thorough examination of failures that we are able to determine their cause. Once that is done the prevention of a repetition of such a failure is principally a matter of intelligent engineering and conscientious production.

It is significant of the care exercised in aircraft manufacture that 90 per cent of the failures which do occur in metal parts happen during test. Had these tests been omitted their study and cure would have been left to chance and to an accumulation of so-called "practice" experience, as is still done in most industries. The number of actual failures in service would have been much greater and the cost in money and in lives vastly increased.

We have another illustration of the indirect influence which aviation is destined to exert on our habits of thought and in the development of our arts and sciences.

Oral Discussion

B. F. SHEPHERD: Mr. Sisco showed a micrograph showing a fillet of improper shape. It seemed from an examination of the photomicrographs at 100 diameters that it was rather a question of an improperly machined fillet rather than an improperly shaped fillet.

A. C. JONES: This interesting paper and the discussion pertaining thereto is significant in view of the great advances that are now being made in the application of the airplane to transportation. Consideration of the fatigue values of steels intended for vital parts, particularly of the engine,

must include the influence of the design as it may affect the internal stresses produced by heat treatment, not found perhaps in the fatigue specimens that were the basis for the endurance limit used. Such stresses may reduce the factor of safety employed.

It would therefore seem of advantage to choose that analysis of steel which would necessitate a higher tempering temperature after an oil or water quench, to give a required tensile and elastic value and thereby reduce, if not eliminating the internal stresses which may affect the endurance limit of the part in question.

MR. HOLDEN: I would like to ask Mr. Sisco if he would tell us what method he has found to be the best means for detecting hairline seams and checks in finished hard parts.

Author's Reply to Discussion

The writer wishes to express his appreciation to Messrs. Downes and Knerr, and to Lieutenant Commander Nelson for their written discussions and to the ones who participated in the oral discussion of the paper.

The discussion by Mr. Downes on streamline wire was especially interesting in that it brought out a fact too little appreciated by users of highly stressed, hard-drawn products; that high internal stresses may be present. The U. S. Army Air Corps has been cognizant of this fact but has never encountered instances where failure of streamline wire could be traced, or attributed, to internal stress. Streamline wire is about the only hard-drawn and heat treated section used in military aircraft construction that is not tempered before use; that is because the practice used in manufacturing this wire is the only one now known that will produce the desired combination of ultimate strength and resistance to bending.

Streamline wire failures in military aircraft are not frequent and when they do occur there are usually a number of failures on some specific type of airplane. Repeated tests have shown that a streamline wire even though hard drawn will not fail under unrestrained vibration. It is only when vibration is restrained that failure occurs. In all cases in our experience when an epidemic of failures of streamline wire has occurred the trouble has been corrected by redesigning the end fitting. An analogous instance of steel of high internal stress is some landing gear axles. These axles are hardened and tempered at 650 degrees Fahr. There are, no doubt, high internal stresses present in this particular member but on the other hand these axles have given ten years of hard and satisfactory service on the old Army DH airplanes with practically no failures.

Lieutenant Commander Nelson's discussion is especially pertinent at the present time when the aircraft industry has finally reached the state of demanding a sizable tonnage of special steels. Needless to say, the inspection of steels by Air Corps inspectors is probably the most rigid possible to give, but in spite of that, as noted in the paper, considerable inherently bad steel, or steel not properly heat treated, or mixed in analysis finds its way into aircraft construction. In the past, due principally to the relative unimportance of aircraft, as an industry, the steel maker has not always offered the airplane

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builder the co-operation necessary to insure a uniform, quality product. The aluminum industry has far surpassed the steel industry in this respect. Things, however, are improving and it probably will not be long before troubles due to bad steel, or improperly heat treated steel will be over.

Mr. Kuerr's discussion emphasizes the importance of the examination of failures and the relatively small number occurring in service compared with those in test. It goes without saying, of course, that we cannot afford to have service failures if it is possible to avoid them and it is possible to reduce service failures to a minimum by rigid and thorough testing. This is one of the chief functions of the Materiel division at Wright Field. Many of our tests are very severe and are designed not only to secure performance data, but also to detect any weaknesses in materials or design.

In answer to Mr. Jones' suggestion for using steels which can be tempered at relatively high temperatures; this has been the policy of Air Corps designers. There is very little high stressed, heat treated steel used in an airplane or engine that is not tempered at 800 degrees Fahr. or above.

The best method we have found for detecting hairline seams in crankshafts is to use a high powered hand magnifier on the polished surface of the bearings. If the number of hairline seams is large a section is cut out of the shaft and coarse etched in hot dilute hydrochloric acid. This etch will exaggerate any small defect and will frequently enlarge very small seams so they can be readily detected. -Fig. 16 of the paper is a section that had been coarse etched for this purpose, followed by buffing with a wire wheel. That illustration shows how these small defects are revealed.

MAGNETIC INVESTIGATIONS OF CARBON STEEL

BY CLIFFORD C. DUELL

Abstract

This paper reports the results obtained by magnetic methods in an investigation of irons and steels of various carbon contents and in various conditions.

It was found that in steels containing cementite a drop in magnetism occurs from 150 to 220 degrees Cent. The author sets forth the hypothesis that carbon is soluble in alpha iron but carbide is not.

In steels containing only martensite decreased magnetism occurs at 300 degrees Cent. Martensite is defined as a crystalloid material which results when a solid solution in gamma iron is cooled at such a rate as to permit the gamma iron to transform to alpha iron, but not to permit the alloying element to form a chemical compound or a solid solution as the case may be.

Cementite is stated to be a stable compound at all temperatures below 745 degrees Cent., and with a proper carburizing heat treatment below this temperature, iron may be converted entirely into carbide.

THIS paper grows out of the research conducted by the author on some of the allotropic properties of iron. There has been some work done on the magnetic properties of iron and steel, and it was thought desirable to make a complete series of tests of these magnetic phenomena for various carbon contents and for various conditions of the material, under carefully controlled conditions, so that the relations between the variables might be found.

Considerable work has been done to account for the phenomena at room temperatures, and on the basis of the observed facts, theories have been set up by Ewing (1), (2), (3),¹ Weiss and Beck (4). Honda (5), by an extension of the Weiss theory of the magneton, was able to set up equations whereby he derived

¹The figures appearing in parentheses refer to the bibliography appended to this paper.

From a thesis by Clifford C. Duell, submitted to the Massachusetts Institute of Technology in partial fulfillment of the requirements for a doctorate degree. The experiments were conducted in Dr. Williams' laboratory. The author is a Lieutenant in the United States Army and is stationed at Schofield Barracks, Honolulu, T. H. Manuscript received September 18, 1928.

the magnetization curve and the hysteresis loop with some degree of success.

The first curve to show the magnetism of a specimen when heated was due to Hopkinson. He found that when the temperature had risen to slightly below A_{c2} a great increase in magnetism took place, and that a specimen of wrought iron was highly susceptible in a weak field as long as the temperature did not exceed 775 degrees Cent. (1427 degrees Fahr.). Roland (6) also found higher susceptibility for weak fields than for strong fields at high temperatures.

Ewing (1) divided the magnetization curve into three parts as the magnetizing force was increased. The first part is characterized by a small change in slope. This is followed by a stage of easily acquired magnetism with a steep slope, and lastly by a second period of small slope as the specimen approaches saturation.

The whole curve becomes reasonable if we visualize a large number of unit magnets due to electron fields. When a magnetizing force is applied, the fields all tend to turn in the direction of the force, and as they turn into the field, if we resolve the unit magnet into two parts, one parallel to the direction of the field and one perpendicular to it, we will find that, assuming the angular rotation to be the same for each increment of magnetizing force, the parallel component increases in proportion to the sine of the angle of rotation, and hence, much more rapidly near the middle of the quadrant than at either end.

What has happened when we subject the specimen to a low field during heating is that the unit fields have rotated as far as they are able to with that particular force applied, and as we raise the temperature, the internal energy increases until finally it becomes such that the unit magnets are able to rotate into the plane of the force, with consequent increase in magnetism.

However, if a medium strong field is applied the magnetism will remain more or less constant until the critical range is reached, and if a strong field is used, the magnetism will decrease even from room temperature.

In the case of a strong field, the specimen is thoroughly saturated at room temperature, and as we increase the energy by heating, the unit magnets are able to swing more and more

away from the impressed force with a gradual decrease in magnetism.

In this investigation the effort was made to eliminate this effect of field strength as much as possible. This is rather an anomaly, since a field must be used to get magnetism. If a small field is used the magnetic curve will be complicated by the increase in magnetism near the critical range, and with a very strong field the curve drops off gradually, possibly masking any points of inflection. It was the aim to use such a value of field strength as will give a horizontal curve for the most part, and hence to eliminate the change in slope due to the field. For most iron and steel, an internal magnetizing force of ten gilberts per centimeter will bring the specimen well up to saturation, and this was the value aimed at.

The effect of the free ends of the specimen must be considered in determining the internal magnetizing force. If a straight bar is placed in a solenoid through which a current is flowing, free poles are set up on the ends of the bar which seriously disturb the uniformity of the field. The effect of these poles, called the demagnetizing factor, was determined by Morton (7), and his results applied to this work give a demagnetizing factor of about 75 per cent. This value has been used throughout.

PREVIOUS WORK

Benedicts and Dearden (8) reported the investigation of magnetic changes in iron and steel at temperatures below 400 degrees Cent. (752 degrees Fahr.) and find weak points of inflection at 130 degrees Cent., and at 210 degrees Cent. in the direction of lowered magnetism. In the hardened state points were found at 120 and 260 degrees Cent. and a gradual bending off of the curve at 330 degrees Cent., indicating loss of magnetism. The present author would point out two faults in their work. In the first place the method used was the magnetometer, which is notoriously sensitive to outside influences, and, secondly, the impressed field was only ten gaussess on a specimen 10 millimeters long by 1 millimeter in diameter. This means that the field was a minute one, and, consequently, brings in the effect of the influence of the field, as pointed out previously.

Honda (9) noted a discontinuity in the magnetism of iron

and steel at low temperature. In investigating twelve samples of varying carbon content he found that a magnetic drop occurred between room temperature and 220 degrees Cent., with a maximum slope at 170 degrees Cent. The amount of the discontinuity present varied with the quantity of cementite present, up to about 0.80 per cent carbon, after which it decreased slightly. There was no corresponding break in the curve for pure iron or for gray cast iron.

The conclusion is reached that since the magnitude of the change is dependent on the quantity of cementite present, that cementite becomes nonmagnetic over this temperature range, and Honda proposes to call this point Ao. That this conception is incorrect will be shown in the progress of this work.

TEST APPARATUS USED

After trying several different arrangements the following equipment was found satisfactory. One hundred turns of No. 16 nickel-chromium wire were wound directly on the surface of a $1\frac{1}{4}$ by 12-inch alundum tube. The turns were spaced uniformly except for a half inch at each end of the tube. This coil was covered with alundum cement and formed both the heating element and the magnetizing coil. A ballistic galvanometer was used, and, in order to eliminate thermoelectric effects, it was found necessary to place several glass rods lengthwise along the heating coil and wind the secondary on them. The secondary consisted of 120 turns of No. 18 copper wire wound on the glass rods, so that it did not touch the furnace at any point. This eliminated the thermal effects in the secondary circuit, since there was some air circulation between the two coils. The secondary was cool to the hand at all times and the drift of the galvanometer was not more than 0.3 centimeter on any run.

A ballistic galvanometer was used to measure the flux. The instrument had a period of 37 seconds, amply large for the size specimens used. A decade resistance box was used in series with the galvanometer, to insure a suitable deflection. Deflections were read on a ground glass scale placed 30 inches in front of the instrument. The zero deflection of the instrument was checked by heating the furnace with no specimen inside to determine the effect of changing current due to increased resistance of the heating

coil. Nickel-chromium wire has a small temperature coefficient, and the deflection dropped off less than 0.1 centimeter as the temperature increased, and since this is less than 0.3 per cent of the total deflection with the specimen inside, it was neglected.

Deflections were obtained by reversing the heating current with a double throw switch, reversal being necessary to eliminate any residual magnetism in the specimen.

The heating coil was connected through the double throw switch and a rheostat to the 115-volt direct current lines. There was some doubt as to whether or not it would be necessary to correct the deflection for the current flowing at the instant of break, and careful measurement showed that the current was constant enough to neglect this factor.

Temperatures were measured by reading, on a potentiometer, the e.m.f. developed by a thermocouple. An iron-constantan couple was used, which was standardized against the platinum, platinum-rhodium laboratory standard over the whole range, and checked for the freezing point of sodium chloride, as a further precaution.

Specimens in all but one or two instances were drilled along their axes, and the couple placed in the center of the specimen. In a few of the specimens, the material was too hard to be drilled, and in these cases the thermocouple was placed in a recess ground in one side of the specimen. In nearly all cases the specimens were $\frac{3}{4}$ -inch round bars, and were seven inches long.

The test specimen and the thermocouple were placed in the furnace and the heating switch closed. Heating was carried on slowly, so that uniform conditions throughout the specimen were obtained. After the magnetic critical range had been passed, the switch was left open, and then closed again as the specimen became cool enough for each succeeding point. Each run required about four hours from room temperature back to the lowest reading taken on cooling.

Thermal e.m.f. was read every 0.5 millivolts over the entire range, and in the areas of rapid change the readings were taken every 0.25 millivolts, about equal to 4 degrees Cent. Smaller temperature intervals were not practicable, as the period of the galvanometer was such that with a smaller interval, the mirror did not come to rest before it became time to take the next reading. Enough resistance was cut into the galvanometer circuit to give

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a desirable deflection for each specimen. The resistance used was different for different specimens, since they varied in permeability and in some cases were of different size. It was found that a current of 6.10 amperes gave a good rate of heating and that value was used throughout.

All magnetic curves were drawn with the millivolts representing temperature as abscissa, and centimeters deflection of the galvanometer as ordinates. The magnetizing force given on each

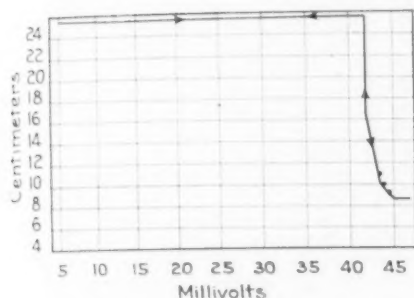


Fig. 1—Magnetization-Temperature Curve for Electrolytic Iron. $H = 7.60$.

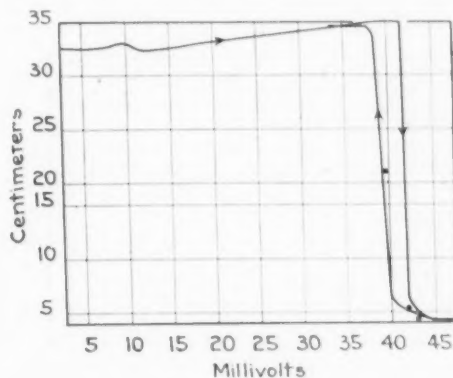


Fig. 2—Magnetization-Temperature Curve for Spheroidized Steel Containing 1.00 to 1.10 Per Cent Carbon. $H = 7.60$.

curve is the net internal magnetizing force after allowing for the demagnetizing factor. It was not thought important to determine relative values of permeability between specimens, so the galvanometer resistance was varied to give a desirable deflection. The essential points are the form of the curves and the temperatures at which the various points of inflection occur. The curves are drawn as continuous lines, the small arrow head showing the direction in which the curve is progressing. The only experimental points shown are those which did not fall on the curve as drawn. Dots represent points on heating, and dots with circles around them represent points on cooling.

THE MAGNETIZATION-TEMPERATURE CURVES

Fig. 1 gives the magnetization-temperature curve for a specimen of commercial electrolytic iron. As had been hoped, the magnetizing force was such as to keep the curve horizontal until the critical range is reached. The drop occurs very suddenly in this material and then shades off to zero at 45 millivolts (778

degrees Cent., 1435 degrees Fahr.). The curve on cooling, except near the foot, was so close to the heating curve that it is impossible to draw two curves. Tests of ingot iron show an almost identical curve. Specimens of electrolytic iron and ingot iron quenched in iced brine from a temperature of 1010 degrees Cent. (1850 degrees Fahr.) show no appreciable difference from the annealed specimens.

The next series was made on 0.10-0.15 per cent carbon steel and gave curves similar to the above. Tests were made of steels

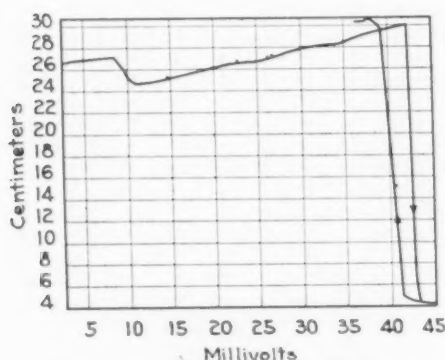


Fig. 3 — Magnetization-Temperature Curve for Steel Cooled in Air from 870 degrees Cent. Carbon Content 1.00 to 1.10 Per Cent. $H = 7.60$.

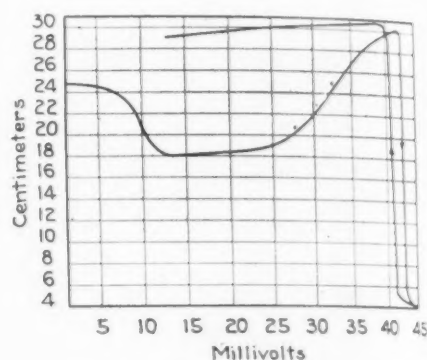


Fig. 4—Magnetization-Temperature Curve for Steel Quenched in Oil from 870 degrees Cent. Carbon Content 1.00 to 1.10 Per Cent. $H = 7.75$.

containing 0.20-0.30 per cent, 0.40-0.50 per cent, 0.80-0.90 per cent, 1.00-1.10 per cent, and 1.30-1.40 per cent carbon. Of these, due to space limitations, only the 1.00-1.10 per cent series is shown. The discussion of the curves applies to all of them, however. The first curve, Fig. 2, shows the material in the spheroidized condition. The curve starts out generally horizontal until about 10 millivolts is reached, where the specimen underwent a decrease in magnetism, not very pronounced but, nevertheless, present. The curve then rises slowly until at an indefinite temperature near the critical range a sudden drop occurs. The material is nonmagnetic at 45 millivolts, but on cooling there is a pronounced lag in regaining its magnetism. The gradual rise in the curve was found to be due to a condition approaching tempering, since it was increased by an increased field and not decreased, as we would think if it were due to the effect of field strength. This rising slope was much more pronounced in the hardened specimens.

Fig. 3 shows the same material after being cooled in air from

870 degrees Cent., resulting in a sorbitic structure. In this condition the point near 10 millivolts is more pronounced, but it starts sooner and ends later than for the spheroidized condition. Otherwise the curve is similar to the preceding.

The same specimen, quenched in oil from 870 degrees Cent., resulted in Fig. 4. The structure was troostite. The general characteristics of the curve are the same, with a still earlier start and later finish of the drop at 10 millivolts.

The specimen was then quenched from 870 degrees Cent. in water and tested with the result shown in Fig. 5. The structure

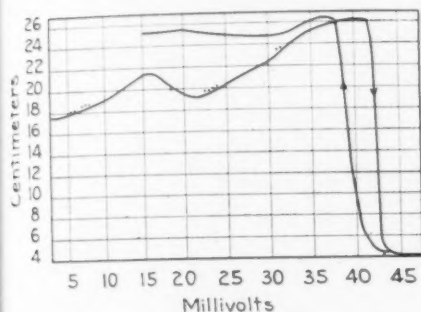


Fig. 5—Magnetization-Temperature Curve for Steel Quenched in Water from 870 degrees Cent. Carbon Content 1.00 to 1.10 Per Cent. $H = 7.75$.

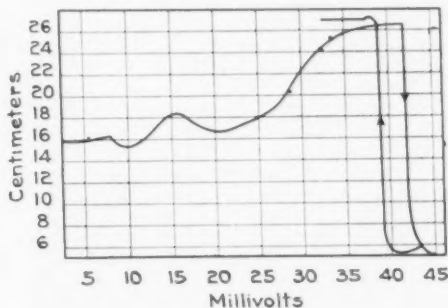


Fig. 6—Magnetization-Temperature Curve for Steel Quenched in Water from 800 degrees Cent. Carbon Content 1.30 to 1.40 Per Cent. $H = 7.60$.

of this specimen is martensite and the resulting curve is entirely different from any of the preceding. First comes a gradual increase until about 15 millivolts at 275 degrees Cent. (527 degrees Fahr.). A more or less extended drop sets in, which is then followed by the usual increase up to the critical range. This point is characteristic of all specimens containing martensite. It is first seen in the 0.40-0.50 per cent carbon specimen. In those specimens which did not contain martensite the drop at 10 millivolts is proportional to the carbon content for the same condition of the material. Specimens which contain both martensite and troostite show both points, as is shown by Fig. 6, which is of 1.30-1.40 per cent carbon, and contained both martensite and carbides after being quenched in water from 800 degrees Cent. (1470 degrees Fahr.). Quenching this material from 1200 degrees Cent. (2190 degrees Fahr.) entirely suppressed the point at 10 millivolts, but resulted in the martensite point being more pronounced than ever.

Further tests were made on malleable cast iron. Fig. 7 shows

the curve for the unannealed material, and shows that the drop at 10 millivolts is much greater, perhaps equal to 40 per cent of the total magnetism. Each succeeding run on the same bar shows a diminution of this drop, probably due to the continued tempering of the bar. It should be noticed that in the steels there is considerable lag in the regaining of magnetism, but very little lag in the case of the malleable iron. Further, this material does not become paramagnetic until a temperature of 815 degrees Cent.

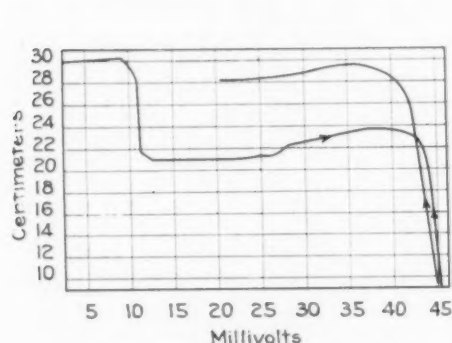


Fig. 7—Magnetization-Temperature Curve for Unannealed Malleable Iron. $H = 7.75$.

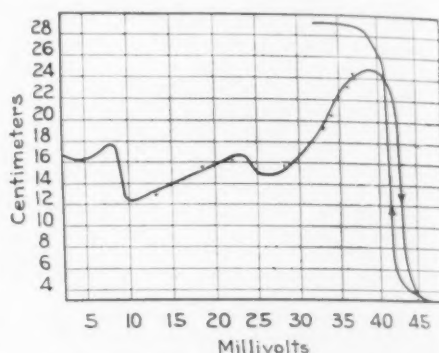


Fig. 8—Magnetization-Temperature Curve for Annealed Malleable Iron, Water Quenched from 1150 degrees Cent. $H = 7.75$.

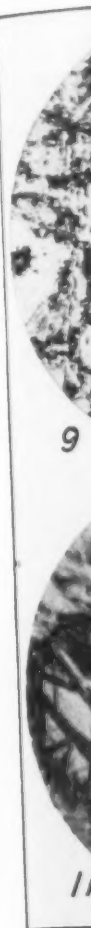
is reached. The bar possesses a strong thermal critical point at 790 degrees Cent., and it may be possible that high silicon has affected the paramagnetic point.

Fig. 8 shows the same characteristics in a well annealed bar of the material which was quenched in water from 1150 degrees Cent. (2100 degrees Fahr.). Both points are well pronounced. The structure of the material is shown in Figs. 9 to 12. The bar in the annealed condition gives the same kind of curve as does pure iron.

DISCUSSION AND CONCLUSIONS

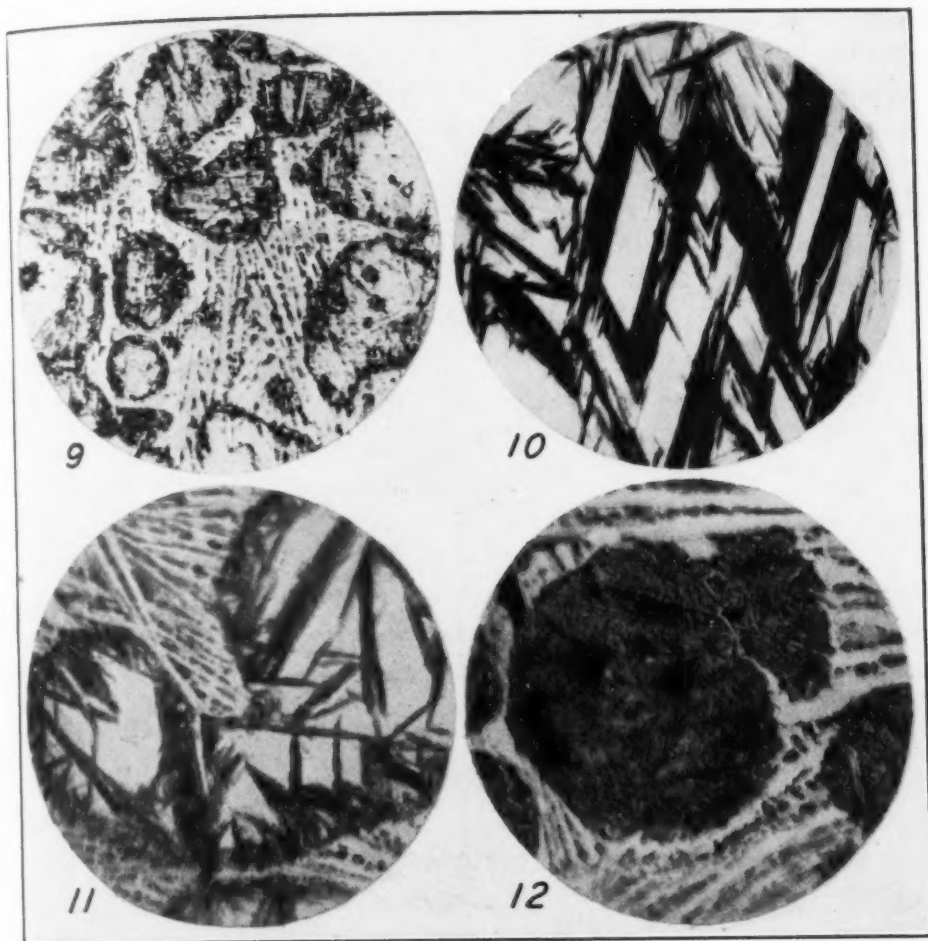
The existence of a magnetic critical point near 200 degrees Cent. in the direction of lowered magnetism was noted previously. This has been thought to be approximately proportional to the quantity of cementite present, and to mark the point where cementite becomes nonmagnetic.

This view is obviously incorrect, since it is shown that the magnitude of the point depends on the state of dispersion of the carbide, being barely visible in the spheroidized condition and amounting to, perhaps, thirty per cent of the total magnetism in



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Photomicrographs of Annealed Malleable Iron, Water Quenched from 1150 degrees Cent. $\times 500$. Fig. 9—As Hardened. Fig. 10—Tempered at 260 degrees Cent. Fig. 11—Tempered at 315 degrees Cent. Fig. 12—Tempered at 370 degrees Cent.

the case of large quantities of troostite, and increasing as we approach this condition. The point is entirely absent in martensite, which does not contain needles, and is present in decreased magnitude in the case of troostite, which does contain quantities of needles.

It is well known that the susceptibility of iron compounds is very small and it is unreasonable to think that cementite should be highly magnetic enough to cause this large drop if it did become nonmagnetic at 200 degrees Cent. Small pieces of relatively pure cementite were manufactured by the method used by Charpy, and were not magnetic enough to be attracted by a magnet. The evidence shows that the magnitude of the drop depends

not on the quantity of cementite present, but on the area of surface of that cementite in contact with the ferrite.

As to the exact position of the point the evidence is not so conclusive. The start of the drop is rather indefinite, and it begins earlier with the more disperse condition of the carbide. The writer has taken the start as 142 degrees Cent. (288 degrees Fahr.). The end is also rather indeterminate, being later for the more disperse condition, but a good value seems to be 210

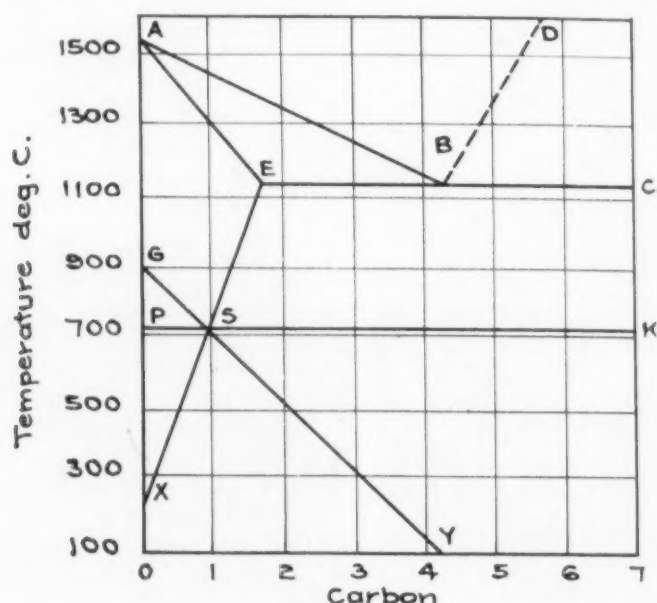


Fig. 13—Modified Iron-Carbon Diagram.

degrees Cent. (410 degrees Fahr.). Honda takes 220 degrees as marking the end of the point. Thompson and Whitehead (10) found two magnetic points in high carbon steels, one at 160 degrees Cent., and the other at 200 degrees Cent. They were unable to decide whether they had observed the beginning and the end of the same phenomenon or marked different ones. It seems that the correct view is that they observed the beginning and end of the same transformation.

MODIFIED IRON-CARBON DIAGRAM

Let us, for the moment, eliminate from the iron-carbon diagram both the A_1 and the A_3 lines, and, further, let us project the cementite line according to the usual principle that the line

will continue in the same direction in the region of instability as in the region of stability. Now there is left a straight case of limited solubility of one element in another, and we get the type diagram shown in Fig. 13. Carbon will continue to separate out along the line ESX until it intersects the zero carbon axis. Note that this intersection comes at 220 degrees Cent. Then here is the point of limit of solubility of carbon in iron, and it is the same point that we have determined magnetically. In other words, in the vicinity of 220 degrees Cent., carbon can go into solution in iron, but carbide cannot.

The modern conception of a eutectic or eutectoid is that there is a region of disorganization several hundred atoms deep at the interface of the two components. When the carbide forms at 743 degrees Cent. (1370 degrees Fahr.) it is extremely probable that there are certain atoms of carbon which are so situated that they could not form complete molecules of carbide, and, consequently, were left hanging in space, as it were. When we heat a specimen through this range these atoms go into solution in the iron in the alpha state, accounting for the decreased magnetism of the material as a whole. When these disorganized atoms have gone into solution, no further drop in magnetism takes place until A_{c1} is reached, showing that the carbide is not soluble in alpha iron.

Let us look at it from another viewpoint. Admitting for the moment that the second point on the magnetic curve (300 degrees Cent.) marks the formation of carbide, and that the carbon in martensite exists as carbon and not carbide, the following condition results. On tempering a steel the magnetism increases slowly, but if the steel is in the martensitic condition, some of the carbon will continue to go into solution until the martensite point is reached, and the magnetism will increase more slowly before reaching this point than after passing it, since after the carbide forms, none of it can go into solution. Inspection of Fig. 8 brings this out clearly, where the slope before passing the 300-degree point is much less than the slope after passing it.

Incipient graphitization, which occurs on annealing white iron below A_{c1} , may be due to this carbon which did not form carbide.

The writer believes that were it not for the complicating effects of the carbide which forms at 743 degrees Cent., the limited solubility curve would represent the true course of events.

THE A_3 LINE

Due to the allotropic transformation from body-centered to face-centered cubic, other complications enter. Of course the present conception is that at 743 degrees Cent. the eutectoid present forms austenite. It seems more in keeping with the behavior of other elements that this is not the case. Alloying elements other than carbon cause the A_3 line to occur at progressively different temperatures, and why should carbon be an exception? It seems reasonable to expect that in the absence of carbide, the continued addition of carbon would cause the A_3 change to occur at progressively lower temperatures along the line GSY of the diagram.

The heating curve of a specimen containing, say, 0.40 per cent carbon, gives an absorption of heat at 743 degrees Cent., thought to be due to transformation of the eutectoid present to austenite. As we raise the temperature, ferrite continues to be absorbed, until at Ac_3 it is *all* in solution. But at Ac_3 we get a distinct absorption of heat, and if the ferrite were all in solution at this temperature, what is there to give the sharp break in the curve? If the above is correct, all that can be expected is a slight change in slope of the curve.

Conversely, on cooling, we get a distinct evolution of heat, where, according to the theory, we have nothing to evolve heat.

It is the author's view that at 743 degrees Cent., all that happens is that the carbide becomes unstable, and breaks up, going into solution in alpha iron. Then at the proper Ac_3 for that carbon content, the whole mass changes to austenite. This conception was first suggested by Sauveur in 1906, and seems to have much merit, particularly in view of the heat effects mentioned.

Note that this conception denies the allotropic nature of A_1 for hypoeutectoid steels, and this would be true for hypereutectoid steels, were it not for the fact that in these steels, when A_1 transformation takes place, the material is already above its normal Ac_3 for that carbon content, and, consequently, Ac_3 takes place concurrently.

With pure iron the change occurs at 900 degrees Cent., and with additions of carbon, at progressively lower temperatures. On passing the eutectoid point the carbon is no longer free to lower

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the A_3 point, since it exists as carbide, which is not soluble in any kind of iron.

This paper, then, takes the position, that cementite is a stable compound at all temperatures below 743 degrees Cent., and at all

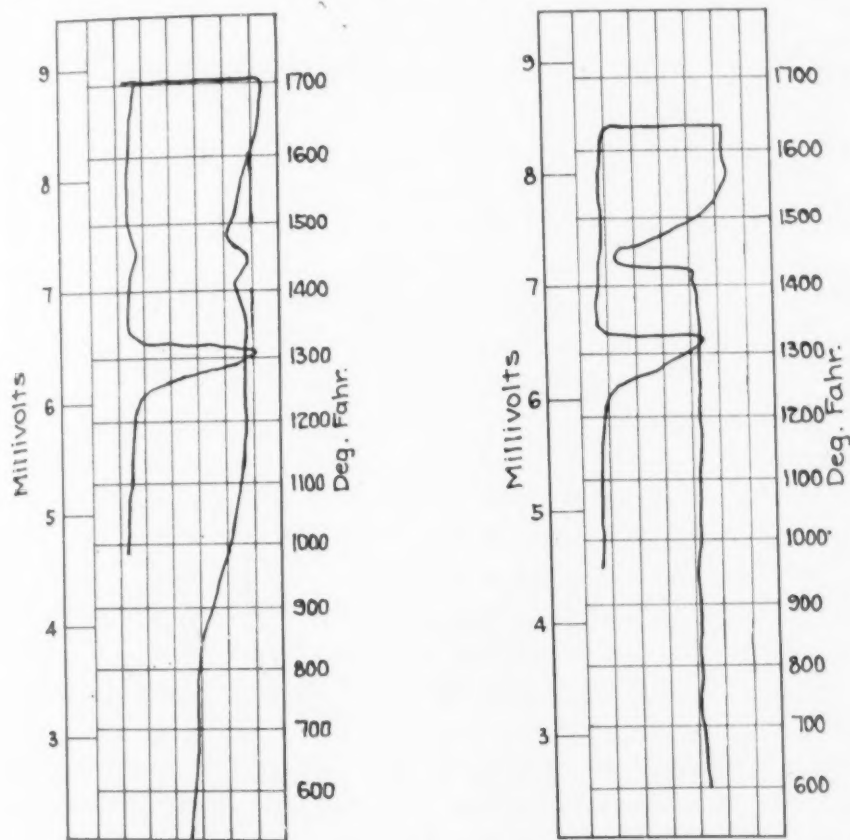


Fig. 14—(Left) Thermal Critical Point Curves of Annealed Malleable Iron. First Run Showing Absence of A_{c1} . Fig. 15—(Right) Thermal Critical Point Curves of Annealed Malleable Iron. Second Run Showing Presence of A_{c1} .

temperatures above 743 degrees Cent., it breaks down into iron and atomic carbon. Given time enough, at a temperature above 743 degrees, this carbon will migrate and form graphite. Atomic carbon is soluble in any kind of iron, but carbide is not.

The objection will be made that by heating a medium carbon steel to a range between A_{c1} and A_{c3} , it may be hardened. This objection is not valid, because if the carbide breaks down and forms a solid solution in alpha iron, this solid solution will be harder than the ferrite surrounding it. The hardening effect of some elements on forming a solid solution is hard to predict, and

it may be possible that the solid solution of carbon in alpha iron is very hard, even though it is not as hard as true martensite.

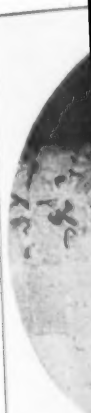
It is true that a change in the solubility of carbon takes place at Ac_1 . Consider the case of malleable iron, well annealed. In this material the iron exists as pure ferrite, and, consequently, should have no critical points except Ac_2 and Ac_3 on heating. Figs. 14 and 15 give the thermal point curve for a specimen of this material. We find no change in slope until the gradual drop in magnetism occurs, with a maximum at 769 degrees Cent. (1415 degrees Fahr.). Next there is another break at 788 degrees Cent. (1450 degrees Fahr.), and that is all.

On cooling there occurs a break at 810 degrees Cent. (1490 degrees Fahr.), which is dragged out in the characteristic manner of the allotropic change. Then at 732 degrees Cent. (1350 degrees Fahr.) comes the sudden evolution due to the formation of carbide.

What has happened is that as the material goes above 743 degrees Cent., the graphitic carbon goes into solution in alpha iron, and as the concentration increases the point is reached where Ac_3 takes place at a temperature lower than 900 degrees Cent. This point is dragged out way up to 900 degrees Cent., due to the varying carbon content in the ferrite. On cooling, the reverse change takes place in the usual manner, with the appearance of Ar_1 , where the carbon which had been in solid solution formed carbide.

It is necessary to enquire why the carbon did not go into solution at a temperature below 743 degrees Cent., since it has been shown probable that the carbon solubility line does not stop abruptly at the eutectoid point. Now if at any stage of the heating below Ac_1 , any carbon should form, it would effectually prevent any further solution of carbon. The process is analogous to the copper plating used to prevent case hardening, only here the coating is cementite, which is stable and insoluble.

It is often stated that no case hardening may be carried out at temperatures below Ac_1 , which means that no pearlite may be formed. It is not true that no carbide may be formed at temperatures below Ac_1 , as cementing has been carried out at temperatures as low as 400 degrees Cent. When the carbide does form at these low temperatures, it is stable and may not be moved except slowly by the process of diffusion.



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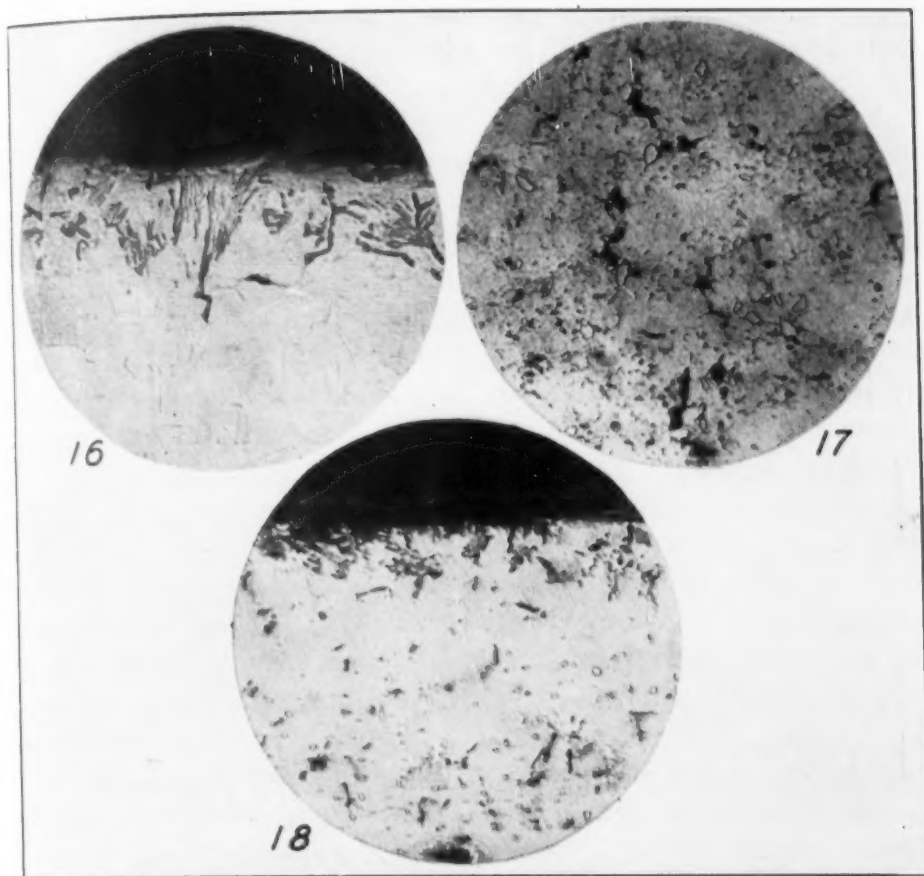


Fig. 16—Photomicrograph of Electrolytic Iron Cemented for 10 Hours at 705 degrees Cent. Etched in Sodium Picrate. $\times 250$. Fig. 17—Pure Cementite. $\times 500$. Fig. 18—Electrolytic Iron Cemented for 10 Hours at 705 degrees Cent., Reheated to 760 degrees Cent. Etched in Sodium Picrate. $\times 250$.

That this is true may be seen from Fig. 16, which is of electrolytic iron, cased in ordinary barium carbonate-charcoal compound for 10 hours at 704 degrees Cent. (1300 degrees Fahr.). Considerable carbide is present, but only one or two grains deep in the material.

PURE CEMENTITE

When iron is exposed to a suitable carburizing agent at temperatures below A_{c1} , carbide is formed, but no solution of carbon in iron. This was shown by Charpy (11), when he found that iron wires and filings, placed in potassium cyanide for 100 hours at 650 degrees Cent., became 100 per cent cementite, the analysis showing 6.72 per cent carbon. If carried out at a temperature

of 1000 degrees Cent. in a current of illuminating gas, the resultant product gave 8.32 per cent carbon, 7.66 per cent of which was in the state of graphite.

Attempts were made to duplicate this on specimens large enough to obtain magnetic curves. Sodium cyanide was used at a temperature of 704 degrees Cent. (1300 degrees Fahr.). The results were only mediocre, as considerable oxidation was brought about by the sodium carbonate, which formed in the long periods necessary. After heating for 144 hours the material consisted of cementite and iron oxide. The wires used broke into too small pieces for magnetic analysis, but, as before stated, it was too paramagnetic to be attracted by a magnet. Fig. 17 shows the resultant product.

It seems probable to the author that carburization can be carried out at low temperatures by carburizing agents which have a relatively stable molecule, but which is small enough to penetrate the iron lattice, such as CO and CN. Some of the molecules will break down and the carbon thereof forms carbide in situ, but no solution had taken place. This is plainly shown in Fig. 18, which is the same specimen as Fig. 16, treated by packing in charcoal and heated to 760 degrees Cent. (1400 degrees Fahr.), and then cooled in air. The carbides are shown more diffused than in the first specimen, but note the nitride needles coming from the other half of the CN radical.

MARTENSITE

The author has read the paper presented by Prof. Sauveur at the February, 1926, meeting of the A. I. M. E., and it is certainly true that the outstanding characteristic of the whole problem is the wide range of opinions ventured. The present author is of the opinion that carbon is not necessary to the formation of martensite.

If we take any material which forms a solid solution with gamma iron, and cool it at such a rate that the gamma-alpha change can take place, but also that sufficient time is not given to the stranger atom to adjust itself to the new condition, the resultant product is martensite. Several instances of this are known. Sykes (12) noted it in the case of iron-tungsten; Bain (13) noted it in the case of iron-chromium, and the present author in the case of iron-nickel. The product has the appearance of marten-

site under the microscope, and this is our usual, and, in some cases, our only means of identifying metallographic constituents.

Martensite is not necessarily hard. It would depend on the size of the stranger atom: an atom nearly the size of the iron atom would give soft martensite and an atom radically different in size would give hard martensite. Tungsten and carbon martensite are hard, but nickel and chromium martensite are soft. In the case of nickel-iron, quenched from above the critical range, there is a distinct change to the martensite structure, but only a small increase in hardness. Further, these stranger elements may form compounds with iron, or they may exist in solid solution, under equilibrium conditions.

The most common conception of the cause of martensite hardness is the fineness of the grain. Undoubtedly this is one of the contributing elements, but it seems probable that strain enters in to a large extent. The X-ray diffraction pattern is clearly alpha iron, but the lines are diffuse. This may be brought about by extremely fine grain, or by warped crystal planes. It is perfectly possible to obtain martensite of such large grains as to be visible to the naked eye, although it is certainly true that most martensite is fine-grained. A combination of the two ideas seems better to fit the facts.

Density considerations lead to the conclusion that in carbon martensite, the carbon atoms occupy the spaces between the iron atoms, and, consequently, it is not a true solid solution. Rather it is *crystalloid*, and the hardness is due both to fine grain and to warped crystal planes.

Therefore the author concludes that martensite is a crystalloid material which results when a solid solution in gamma iron is cooled at such a rate as to permit the gamma iron to transform to alpha iron, but not to permit the alloying element to form a chemical compound or a solid solution in alpha iron, as the case may be.

The author agrees that the so-called martensite needles are not martensite at all, but are troostite. The needles occur at the austenite grain boundaries and it is possible, that on quenching, some of the austenite, when the allotropic change took place, was so situated that the carbon was able to form troostite, that is, the formation of carbide took place.

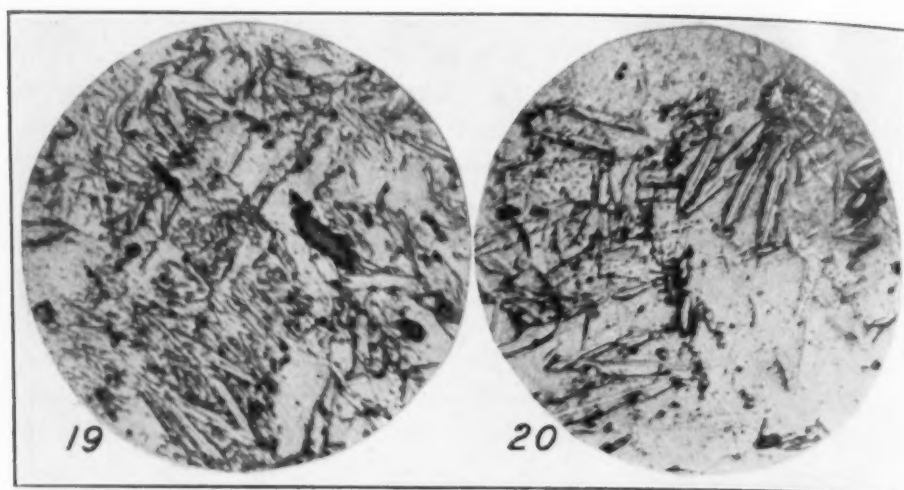


Fig. 19—Photomicrograph of Martensite Needles in Electrolytic Iron when Cemented Below A_{c1} . $\times 500$. Fig. 20—Martensite Needles in Electrolytic Iron when Cemented Below A_{c1} . $\times 500$.

It is perfectly possible to get martensite without these needles and it is possible to get the needles without the formation of martensite. Figs. 19 and 20 are of the iron specimens which were heated in sodium cyanide, and at the end of fifty hours were cooled freely in air. The structure shows many needles, exactly similar to those found in martensite. This undoubtedly shows the presence of carbide in the needles. We must conclude, therefore, that carbon martensite, often, and usually, does contain considerable troostite in the form of needles. In the specimens of malleable iron quenched from a high temperature, many needles are present, and note that the magnetic curves for these specimens containing the needles also show the cementite point. This is added evidence of the troostitic nature of the needles.

The foregoing conception of what happens on hardening is clearly demonstrated in the split A_r transformation. It is true for all alloy steels and for straight carbon steels when the rate of cooling is such that carbide formation occurs first, thereby permitting the allotropic change, only one point is seen, since no carbon is in solution. When cooled more rapidly, or when more of the special element is added, the split transformation occurs and some martensite results. When the rate is such that no carbide forms, but the allotropic change is permitted complete martensitization results and only one point is seen. If both points are suppressed we get austenite.

The Ar'' point should be lower than the Ar' , since the two points are due to different causes. The Ar' point marks the formation of carbide, and the Ar'' point is the allotropic change. If the formation of carbide is suppressed, the carbon will remain in solution, and will depress the Ar'' point along the A_3 line of the iron-carbon diagram, resulting in martensite when the austenite does transform to alpha iron. In other words, the Ar' point is really Ar_1 , and the Ar'' point is really Ar_3 .

THE MAGNETIC MARTENSITE POINT

It appears to the author that the evidence is conclusive that there is no cementite in true martensite. To date the evidence had been more or less indirect. From our magnetic curves, it would follow that if there is cementite in a finely dispersed state in martensite, the magnitude of the cementite point on the curves should be greater than in the case of troostite. The fact of the matter is that the point is gone entirely, but the initial magnetism is much less than in the case of troostite, an indication of solid solution or quasi solid solution.

As those specimens containing martensite are heated to about 300 degrees Cent., a discontinuous drop in magnetism occurs, which may be explained by the formation of iron carbide, the carbon of which was formerly distributed throughout the mass. The exact position of the point is rather changeable, varying from 279 degrees Cent. (533 degrees Fahr.) to 369 degrees Cent. (695 degrees Fahr.), with 300 degrees Cent. as a fair average value. Comparison of the curves with photomicrographs of the hardened specimens show that the point marks the formation of troostite from martensite.

No evidence was found in the specimens that this point marks the formation of troostite from austenite. This probably occurs at a lower temperature and is probably not a very abrupt change. Any indication of this change would be in the direction of increased magnetism. Some of the higher carbon steels, and the malleable iron in particular, containing large quantities of austenite, but no critical points corresponding to its breakdown were found on the magnetic curves. The tempering process, except at the cementite and martensite points, where they exist, is characterized by a gradual increase in magnetism.

The tempering process as shown by the magnetic curves is very similar to that found by Fink and Campbell (14) in investigating tempering with the X-ray. As the tempering process continued, more and more of the iron assumed the dimensions of the alpha lattice, and in this work more and more of the iron seems to become fully magnetic as the temperature is raised.

Some very interesting magnetic curves were obtained with alloy steels which bear out the conclusions above, but space is not available to discuss them.

SUMMARY OF RESULTS

1. On heating steels containing cementite, a drop in magnetism occurs in the range from 150 to 220 degrees Cent. Both ends of the range are indefinite, the drop being more pronounced, and extending over a wider range, as the dispersion of the carbide increases. The hypothesis is set up that carbon is soluble in alpha iron, but carbide is not, and that the drop in magnetism is due to certain atoms of carbon, which did not form carbide, going into solution at this temperature.

2. In steels containing only martensite, the previous point is missing, but a new one occurs at 300 degrees Cent. This is in the direction of decreased magnetism, and corresponds to the formation of iron carbide, coincident with the formation of troostite from martensite.

3. Martensite is defined as a crystalloid material which results when a solid solution in gamma iron is cooled at such a rate as to permit the gamma iron to transform to alpha iron, but not to permit the alloying element to form a chemical compound or a solid solution as the case may be.

4. Cementite is a stable compound at all temperatures below 743 degrees Cent. (1370 degrees Fahr.), and with proper heat treatment below this temperature, iron may be converted entirely into carbide.

The author admits that much of the above is somewhat radical, but it is all based on an effort to make carbon behave the same as other elements do, in its relation to iron. It further seems that the time has come to change the nomenclature of the critical points. They do not always occur in the order $Ac_{1,2,3}$, or $Ar_{3,2,1}$, but may occur in any order. Some better method of identification is nec-

essary, and the author suggests A_{c_a} and A_{r_a} for the body-centered-face-centered change, (a for allotropic); A_m (m from magnetic) for the magnetic change, which is perfectly reversible, and, consequently, needs only one sub-script; and A_{c_p} and A_{r_p} (p from pearlite) for the present A_1 . This nomenclature will effectually designate what occurs at each temperature, and will eliminate ambiguity. The method is suggested as a means of eliminating misunderstanding.

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THE APPLICATION OF SCIENCE TO THE STEEL INDUSTRY—SECTION II

BY DR. W. H. HATFIELD

Abstract

Since the reliability of finished steel products is so dependent on the characteristics of the ingot from which they are produced, this entire section is devoted to a consideration of ingot making. Reports of investigations of ingots made by the Iron and Steel Institute Committee are given. Diagrams are shown of ingot molds of present British design. Photomacrophs and sulphur prints of sectioned ingots are given.

INGOT MAKING

THE characteristics of the ingot determine to a large degree the homogeneity and reliability of the ultimate product. The liquid steel is poured into the mold and freezes. It is indicative of the present state of knowledge that simple though the mechanism of the freezing may appear, yet it is a fact that the knowledge of the operation of physical laws and of the modification of the physical characteristics of iron and its alloys at high temperatures, are not sufficiently complete to enable a final and satisfactory picture to be drawn of the actual process of the freezing of the ingot. Until such an understanding is obtained, technical achievement, as regards the quality of ingot, is limited. This position of affairs was realized in Britain some few years ago, and an influential Iron and Steel Institute Committee is at work composed of representatives of the steel industry, of science and of the services. Two reports have already been published, to which reference should be made.⁷⁵ Some thirty-two steel ingots, varying in weight from 14 cwts. to 172 tons, have been studied. These ingots have been sectioned, and the nature of the distribution of the various elements determined,

⁷⁵Report of the Committee on Heterogeneity of Steel Ingots, *Journal*, Iron and Steel Institute, 1926 and 1928.

This paper is the second section of the third Edward De Mille Campbell Memorial Lecture presented before the tenth annual convention of the society; held in Philadelphia, October 8 to 12, 1928, which is being published in six sections in consecutive issues of TRANSACTIONS. The author, Dr. W. H. Hatfield, is director of the Brown-Firth Research Laboratories, Sheffield, England.

the degree and form of segregation, and the form of internal cavity, whether due to shrinkage or to the evolution of gas disclosed.

An adequate knowledge of the mechanism of the extraction of heat from the freezing steel by and through the mold cannot be properly understood until the physical constants of the mold material, bearing upon the matter, have been determined for the temperatures involved. Enough has already been said to indicate clearly how extremely complex are the phenomena with which one has to deal in endeavoring to provide a reliable picture of the freezing of the ingot; and it is clear, therefore, that both the design of the ingot and of the mold has been heretofore, and is, indeed, even at the present time, definitely empirical. However, much progress has been made, and this can best be exemplified by examples drawn from the reports of the committee referred to.

Fig. 2 is a photograph of a carbon steel ingot made many years ago, which was studied by the Committee as being representative of an extremely large tonnage of the ingots which used to be produced prior to the increased appreciation of the effect of design of mold upon the characteristics of the interior of the ingot. It was cast from the bottom in an octagonal chill mold tapering from $14\frac{3}{4}$ inches across the flats at the bottom, to $10\frac{1}{2}$ inches across the flats at the top. The mold was close-topped and for the greater part of its length was 3 inches thick.⁷⁶ The weight of the ingot was 23 cwts. It will be seen that the major portion of the cavity is contained in a space within a foot of the top of the ingot, but that small shrinkage cavities extend down the whole length to within 12 inches from the bottom. As regards the production of ingots from properly "killed" or piping steel, the Committee found that ingots are now being almost invariably produced with the broad end at the top, the narrow end at the bottom, with adequate arrangements for refractory-lined feeder heads for ensuring the feeding of the ingot proper, free from defects.

Solid carbon and alloy steel ingots produced from "piping" steel, and non-solid ingots typical of the large tonnages of mild steel, have been studied. Typical examples of this work will shortly be given, but the author would here like to indicate some of the encountered difficulties in the way of the solution of the general problem, particularly from the quantitative point of view. A

⁷⁶Example 32, Report of the Committee on Heterogeneity of Steel Ingots, *Journal, Iron and Steel Institute*, May, 1928.

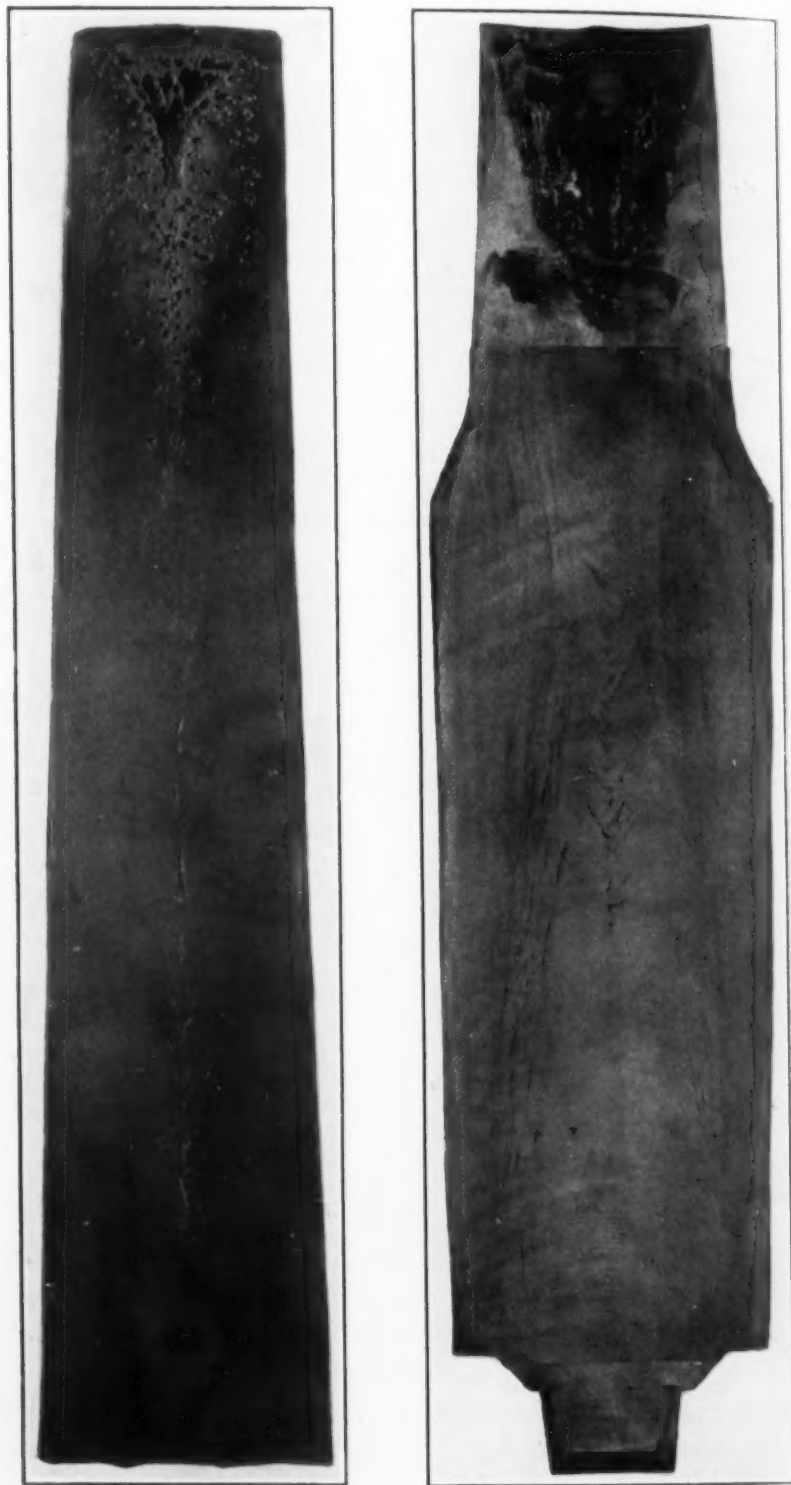


Fig. 2 (left)—Photograph of a Carbon Steel Ingot Representative of Ingots Produced Prior to the Increased Appreciation of the Effect of Design of Molds Upon Characteristics of the Interior of the Ingot. Fig. 3 (right)—Photograph of a Typical Medium Sized Ingot Studied by the Committee. This Ingot was Cast in a Chill Octagonal Mold.

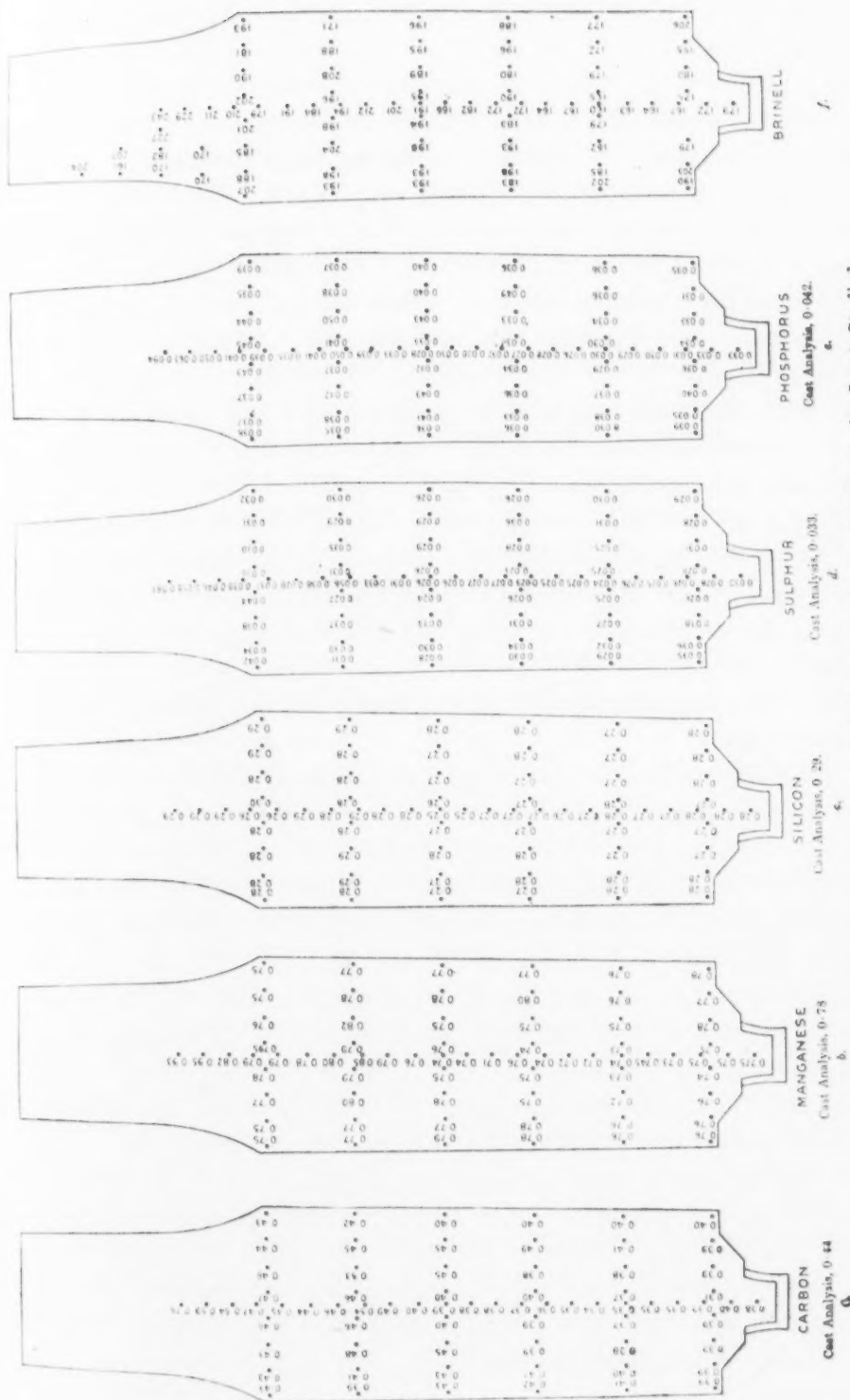


Fig. 4—Illustration Showing Chemical Analyses at Various Points on the Ingot Studied.

typical instance may be given in that it was necessary to explain the pure zone of steel occurring toward the bottom of the ingot along the central axis. Various explanations had been put forward; one school of thought considered that as a result of differential freezing, crystallites of steel, purer than the remaining liquid steel, fell and accumulated toward the bottom of the ingot. Careful consideration of data extant clearly showed that although it had long been held that the change over from the liquid to the solid state was accompanied by a definite increase in density, yet such a view is at the moment only surmise. Researches are being conducted both by Dr. Benedicks in Stockholm, and by Professor Desch in Sheffield, to settle the matter. Another school of thought based its explanation upon the effect of the temperature gradient upon the concentration of the solute in the liquid steel from the central axis to the freezing wall, i.e., the Soret effect; but our knowledge of the quantitative value of this effect in liquid steel is unknown, whilst, incidentally, the experimental method of determining the temperature gradient still remains to be worked out; and it is required to devise some means of providing the data for plotting the increase in thickness of the freezing walls of the ingot against time.

The change in the viscosity of liquid steel with falling temperature is not known. The liquidus and solidus of each type of commercial steel, as a result of the committee's investigations, has for some time been the subject of research under the direction of Professor Andrew of Glasgow. The effect of differential freezing cannot be envisaged in the absence of such data.

As typical of medium sized ingots of this class of steel studied by the Committee, a 25-ton ingot dealt with in the First Report can be instanced.⁷⁷ (Figs. 3 and 4). This ingot was cast in a chill octagonal mold, the weight being 24 tons 17 cwt., and it measured 43 inches at the top of the chill and 40 inches at the bottom. In considering the results of the investigation of this ingot, it is interesting to be able to state that information was supplied to the Committee which is very helpful. The steel was made in the acid open-hearth and the SiO_2 content of the slag was 56.5 per cent at the time the additions of manganese and silicon were made in the furnace. No aluminum was used. The steel on leaving the fur-

⁷⁷Example 12, First Report of the Committee on Heterogeneity of Steel Ingots, *Journal, Iron and Steel Institute*, 1926.

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nace, had a temperature of 2894 degrees Fahr. (1590 degrees Cent.) and at the time of teeming into the mold had a temperature of about 2804 to 2813 degrees Fahr. (1540 to 1545 degrees Cent.). Eighteen minutes were taken in filling the mold, the average speed being at the rate of 1.393 tons per minute. The ingot was sectioned along its central axis and the section thus obtained was ground and polished. In this condition the chill portion of the ingot was free from shrinkage cavity, with the exception of perhaps a little looseness of structure coincident with the upper portion of the axis. As a result of etching with 5 per cent nitric acid in water, the nature and orientation of the segregate was disclosed as illustrated in Fig. 3. This section was subsequently drilled at various positions and the composition determined as disclosed in Fig. 4. Here, therefore, is an indication of the type of heterogeneity which exists in a 25-ton ingot when cast under what must be considered extremely satisfactory conditions from the standpoint of our present knowledge of the physical chemistry of the acid open-hearth process.

In Figs. 5 (a) and (b) and 6 (a) and (b) will be found illustrations relating to two very interesting ingots cast from nickel-chromium steel.⁷⁸

They were intentionally prepared from the same electric furnace heat. The cast weighed 8 tons 10 cwt., and four ingots were cast, all in molds of the same design. They were numbered A, B, C and D in the order of teeming; ingot B was selected as Example 19 and Ingot D as Example 20. At the moment of tapping, the steel and slag analyzed as follows:—

Steel	Per Cent	Slag	Per Cent
Carbon	0.29	SiO ₂	20.74
Manganese	0.42	FeO	2.16
Silicon	0.27	Al ₂ O ₃	2.24
Sulphur	0.016	MnO	0.13
Phosphorus	0.015	CaO	61.42
Nickel	4.38	Cr ₂ O ₃	
Chromium	1.41	MgO	8.14

A reducing slag had operated for some time prior to finishing. The time taken to pour the steel from the furnace was 2 minutes 20 seconds. After 2 minutes 27 seconds had elapsed, the Ingot A was teemed over a period of 3 minutes 40 seconds with a

⁷⁸Examples 19 and 20, Second Report of the Committee on Heterogeneity of Steel Ingots, *Journal, Iron and Steel Institute*, May, 1928.

$\frac{3}{4}$ -inch nozzle. After a lapse of 18 seconds, ingot B was teemed in 3 minutes 40 seconds with a $\frac{3}{4}$ -inch nozzle. 25 seconds later the teeming of ingot C was commenced, and was finished in 3 minutes 55 seconds. A special double nozzle was being employed, and the last ingot D was therefore teemed, after a lapse of 50

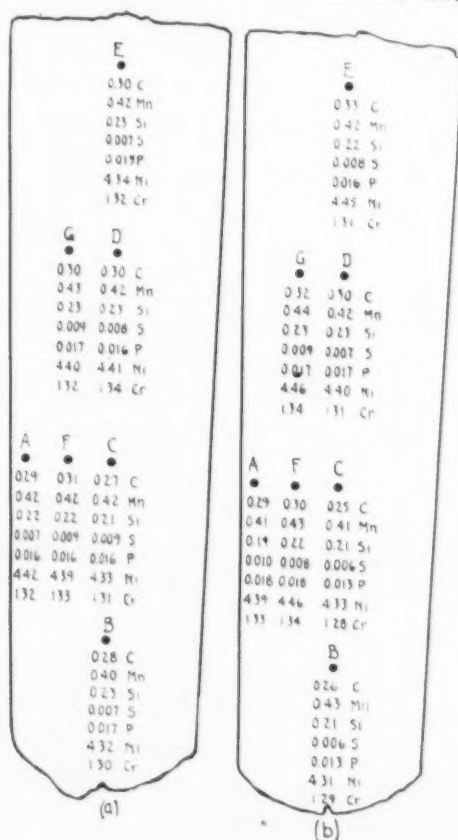


Fig. 5—Diagram Showing Chemical Analyses at Various Points on Two Nickel-Chromium Steel Ingots.

seconds, with a 1-inch nozzle over a period of 2 minutes 10 seconds. The temperature of the steel during tapping into the ladle was determined as 2903 degrees Fahr. (1595 degrees Cent.). The average temperature during the teeming of ingot B was 2894 degrees Fahr. (1590 degrees Cent.), whilst that of ingot D was 2822 degrees Fahr. (1550 degrees Cent.). It will thus be seen that ingot B was cast much hotter and with a smaller nozzle than ingot D, and as these were the only variables in regard to the two ingots, a comparative study is of interest. Owing to the composition of the steel resulting in pronounced air-hardening charac-

teristics, it was necessary to anneal the ingots before sectioning along the axis, and this was done by heating to 1562 degrees Fahr. (850 degrees Cent.), and then very slowly cooling down in the furnace. The feeder heads were then removed. After longitudinal sections had been prepared along the axis, exposing the exact central planes of the ingots, the surfaces were ground and polished and then etched with 2 per cent nitric acid in water. The macrostructures thus produced lent themselves indifferently to reproduction, so it was decided to reproduce the comparative crystal structures of the two sections. This was done by laboriously tracing from the macrophotographs the boundaries of the primary crystals, with the results shown in Figs. 6 (a) and (b). It was hoped that this comparative study would throw some light upon the mechanism of freezing.

It will be seen that in each ingot there is a well-defined outer layer of the first metal to freeze, consisting of very small crystals. Growing from this layer, in ingot B is the region of columnar crystals, which tapers off towards the top of the ingot. These columnar crystals are absent in ingot D. Two important central zones occur in both ingots, a zone of uniformly small crystals in the lower portion, and another of very large crystals in the upper portion. In the ingot D, cast at the lower temperature, the small crystals extend much higher up the ingot, and the zone of large crystals becomes much smaller, as, indeed, do the crystals themselves. It is appreciated that the crystals disclosed are probably pseudomorphs of the original crystals. Even the latter statement tends to be controversial, and it was not the intention of the Committee to discuss the theoretical implications of the data here disclosed. It was suggested that the outer shell of small crystals might be due to modifications resulting from the annealing process. Another ingot in the "as cast" state, and of the following analysis, was therefore selected:

	Per Cent
Carbon	0.31
Manganese	0.46
Silicon	0.27
Sulphur	0.014
Phosphorus	0.020
Nickel	4.19
Chromium	1.29

The ingot was of square section, 12 inches square at the top of the

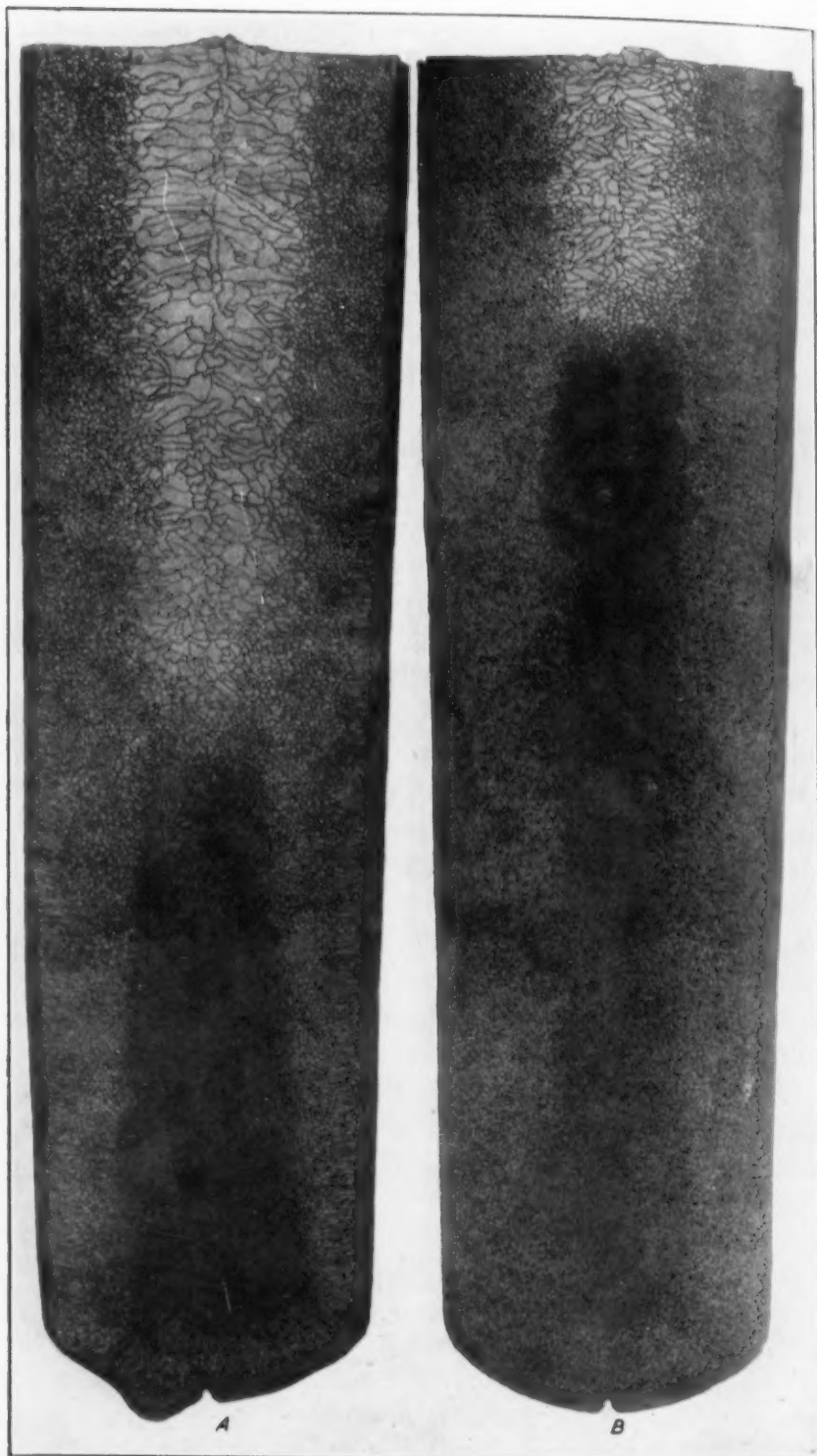


Fig. 6—Photograph Showing Primary Crystals as They Were Traced from the Photomicrographs.



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Fig. 7—Photograph of a Portion of the Fracture Made Across the Middle of the Ingot.

chill portion, and 10 inches square at the bottom; it was cast with a refractory-lined feeder head. This ingot was fractured cold, as shown in Figure 7, which is a large scale reproduction of a portion of the fracture made across the middle of the ingot. It will be seen that the existence of the outer zone of small crystals is confirmed, and that growing from these there is a layer of the columnar crystals. As a matter of interest one of the middle pieces of the ingot was used for further study. This piece of ingot was next annealed at a temperature of 1202 degrees Fahr. (650 degrees Cent.), and a transverse fracture obtained from the middle of the piece, with the resulting structure shown in Fig. 8. From this it will be seen that it is the primary crystallization which is still apparently responsible for the form of fracture.

These examples suffice to give an indication of the characteristics of carbon steel and alloy steel ingots, when produced from the type of steel which freezes solid, so essential in the production of highly stressed forgings and similar steel products.

In Figs. 9 to 12 two examples of ingots are quoted from the Second Report of the Committee, which illustrate the characteristics of the interiors of ingots representative of the very large

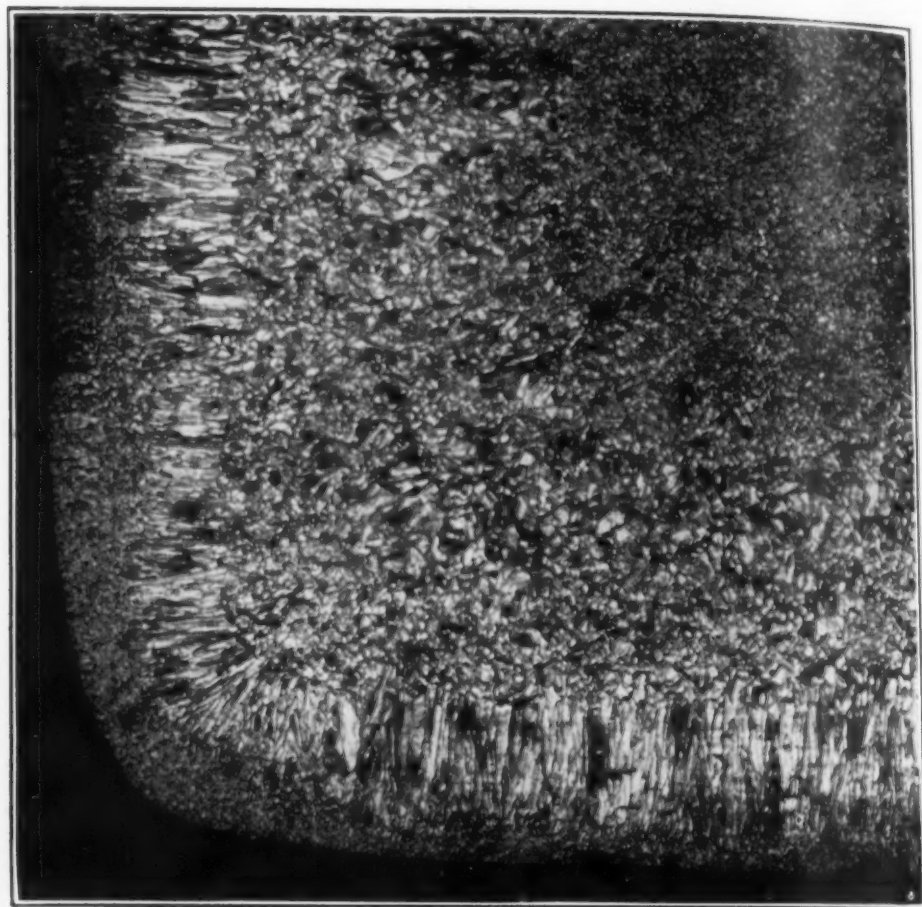


Fig. 8—Photograph of the Transverse Fracture of the Piece of Ingot After Annealing at a Temperature of 1202 Degrees Fahr. (650 Degrees Cent.).

tonnage of mild steel material used for general purposes. The ingot illustrated in Figs. 9 and 10 is an ingot of mild basic steel⁷⁹ of what is known as the rimming type. This steel on leaving the furnace, had the high temperature of about 2975 degrees Fahr. (1635 degrees Cent.). It will be noticed that no attempt is made to feed this type of ingot solid, and that evolution of gas during freezing is largely responsible for preventing the formation of pipe. The macro-etching of the section of the ingot in Fig. 9 (a) clearly indicates the distribution of the gas holes, and quite clearly, in steel of this type, owing to the high temperature at which it is worked, and the carbon content, these holes are expected to weld up. The sulphur print shown in Fig. 9 (b) discloses the interest-

⁷⁹Example 24, Second Report of the Committee on Heterogeneity of Steel Ingots, *Journal*, Iron and Steel Institute, May, 1928.

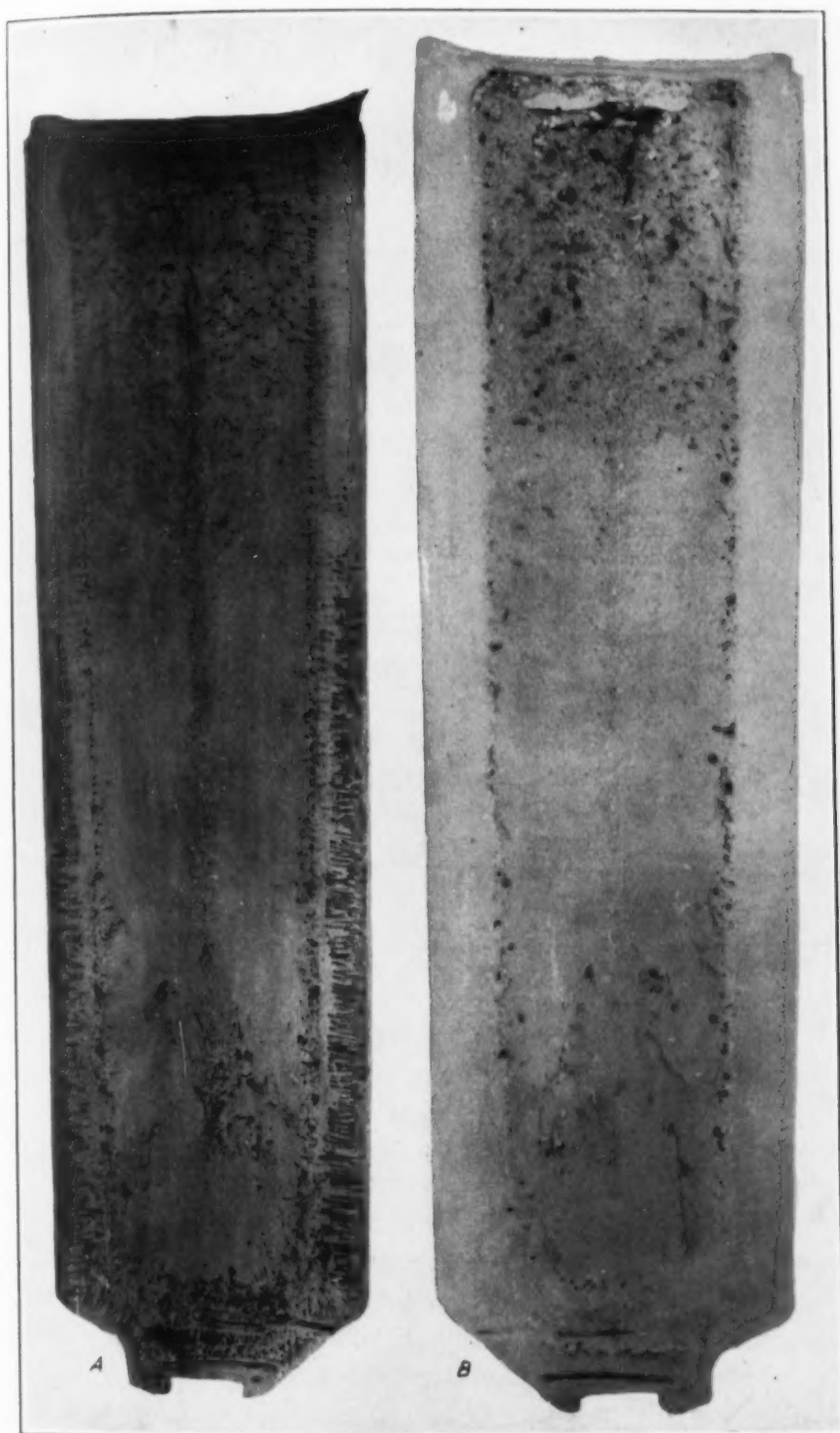


Fig. 9—(a) Photomicrograph of a Section of the Ingot Described Showing Distribution of Gas Holes. (b) Sulphur Print Showing That There is a Definite Zone Containing Lower Sulphur Content Than the Remainder of the Ingot.

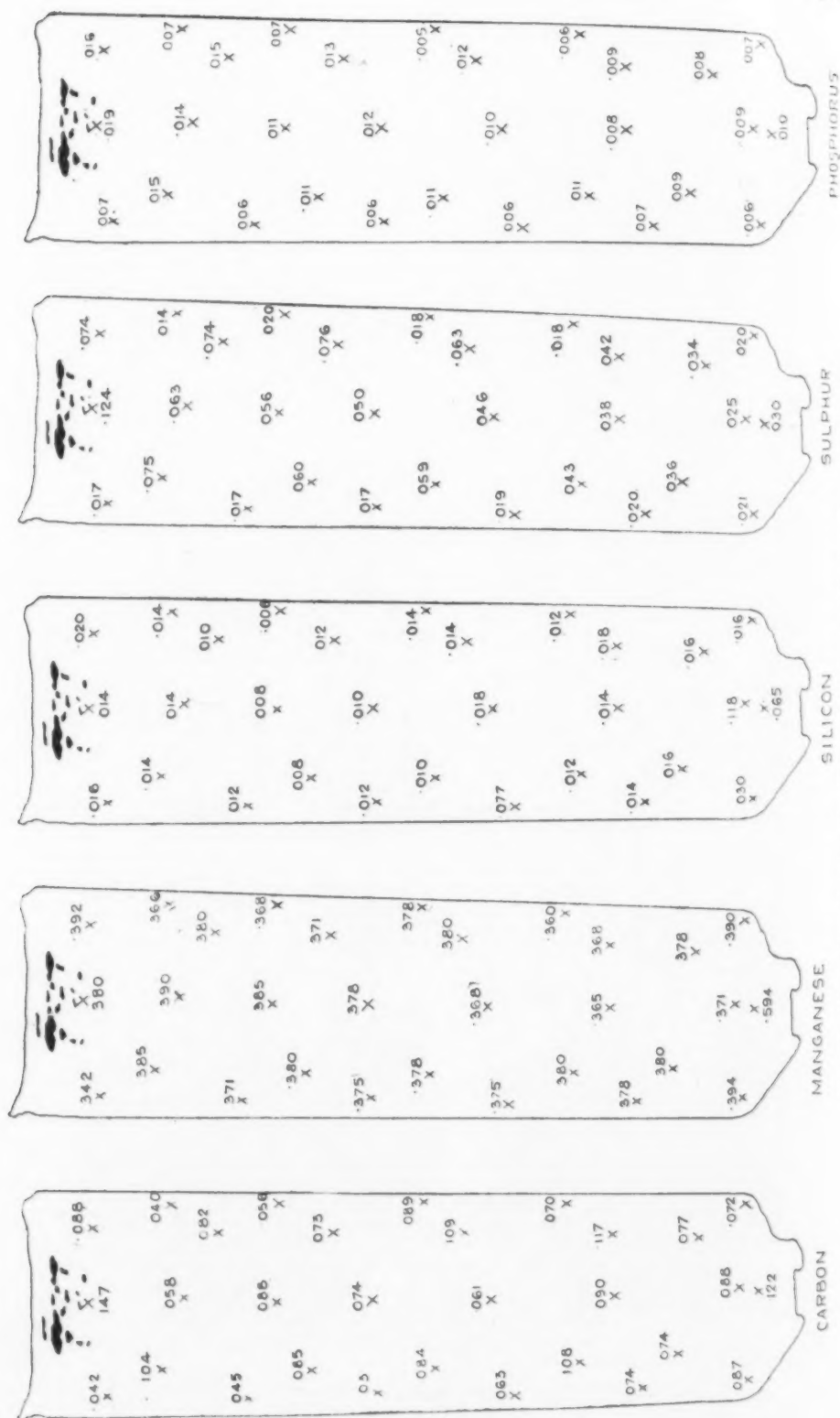


Fig. 10—Illustration Showing the Chemical Analyses at Various Points on the Ingot.

ing feature that there is a definite thick external zone of the ingot, in which the sulphur content is at a lower value than in the remaining portions of the ingot and this is confirmed in Fig. 10, also as regards the carbon and phosphorus contents. This particular ingot weighed 3 tons 1 cwt. The other example,⁸⁰ Figs. 11 and 12, is an ingot weighing 6 tons 15 cwt., of low carbon acid Siemens plate steel and here, in large measure, the same remarks apply.

As regards this type of steel which, it must be confessed, represents the bulk of the mild steel produced in all countries, whilst one cannot but appreciate the fact that owing to the excellent welding properties of such low carbon material and the economic considerations, the steel in the rolled condition with its high yield of the finished product, proves good enough for its purpose, yet one cannot but have at the back of one's mind, the hope that it may ultimately be found practicable to produce even this type of steel in ingots of the same high standard of characteristics as is found in the best type of forged ingots.

Generally reviewing the characteristics of ingots disclosed by the work of our ingot Committee, it may be said that when steel freezes it does so by a process of differential freezing. Consequently, no ingot is uniform in composition. This heterogeneity as regards composition is definitely found to increase with the mass of the ingot, so much so that in some of the largest ingots studied by the Committee, ingots weighing from 100 to 172 tons each, it can safely be said that the variation particularly as regards the carbon content, is so considerable as to necessitate the designing engineer taking into consideration the effect of the variation of the carbon upon the strength of the material. The sulphur and phosphorus were also found to vary within wide limits. As regards the special elements, it was found that nickel remained fairly constant through the mass of the ingot, chromium showed definite evidence of segregation, whilst molybdenum, where studied, showed a very definite tendency to segregate. However, without running to greater length, the author feels that he cannot do better than recommend anyone, interested in steel ingots, to read the Reports of the Committee, in which will be found, carefully stated, the results of the examination of some 32 ingots of varying types and sizes.

⁸⁰Example 27, Second Report of the Committee on Heterogeneity of Steel Ingots, *Journal, Iron and Steel Institute*, May, 1928.

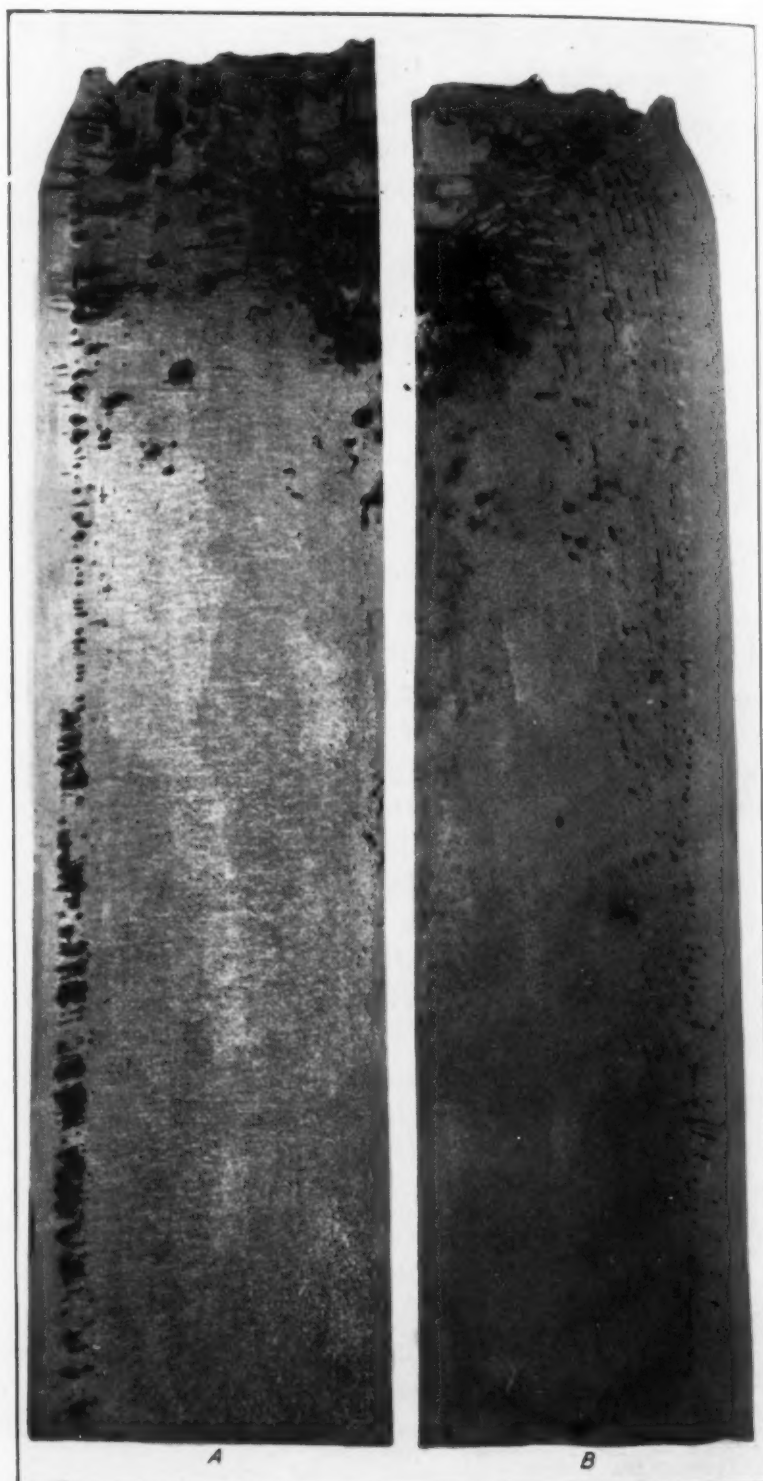


Fig. 11—(a) Photomicrograph of a Section of the Ingot Described.
(b) Sulphur Print of the Ingot Described.

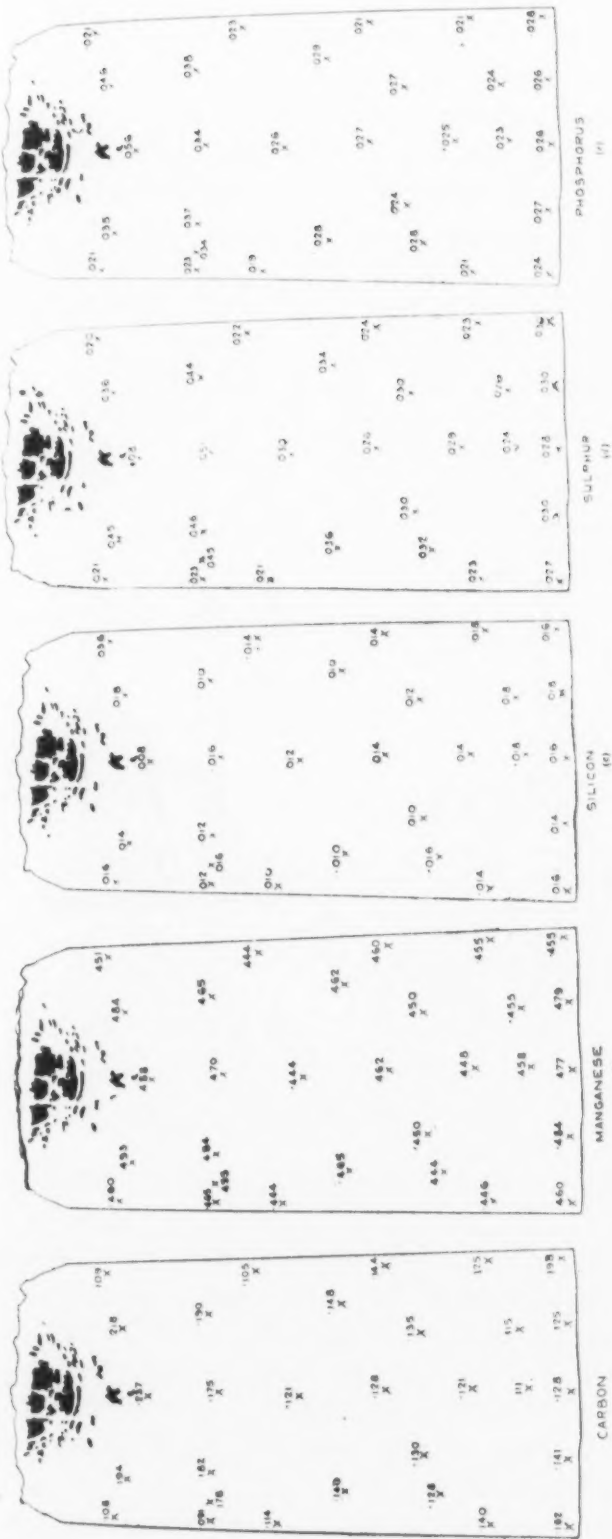
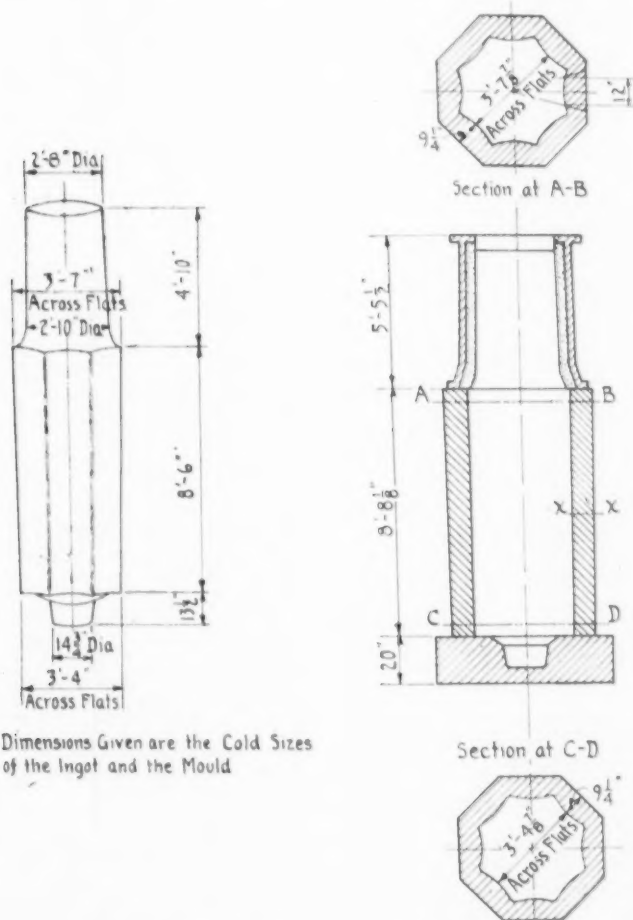


Fig. 12—Illustration Showing Chemical Analyses at Various Points on the Ingot Described.

The Committee has decided not to attempt, for the time being, a picture of the mechanism of freezing, and the author considers he will be wise if he adopts the same attitude.

One point can, however, be touched upon, and that is the undoubted fact that the higher the proportion of nonmetallic



Dimensions Given are the Cold Sizes of the Ingot and the Mould

Fig. 13.—Diagram of Ingot and Mold of Present British Design.

inclusions in the steel, i. e., sulphide, silicates and oxides, the more serious do the results of local segregation, i. e., ghosts, become. It is undoubtedly a fact that particularly when casting large masses, the sulphur should be kept as low as possible, and the steel should be produced by one of the processes which enables the deoxidation to take place in the steel making furnace to the maximum degree possible.

The casting temperature and speed of teeming should be

thoroughly explored as regards any particular type of ingot, and once the best conditions are established, should be rigidly adhered to.

As regards the design of mold, it may be said that with regard to the type of ingot necessary for forgings and high class work,

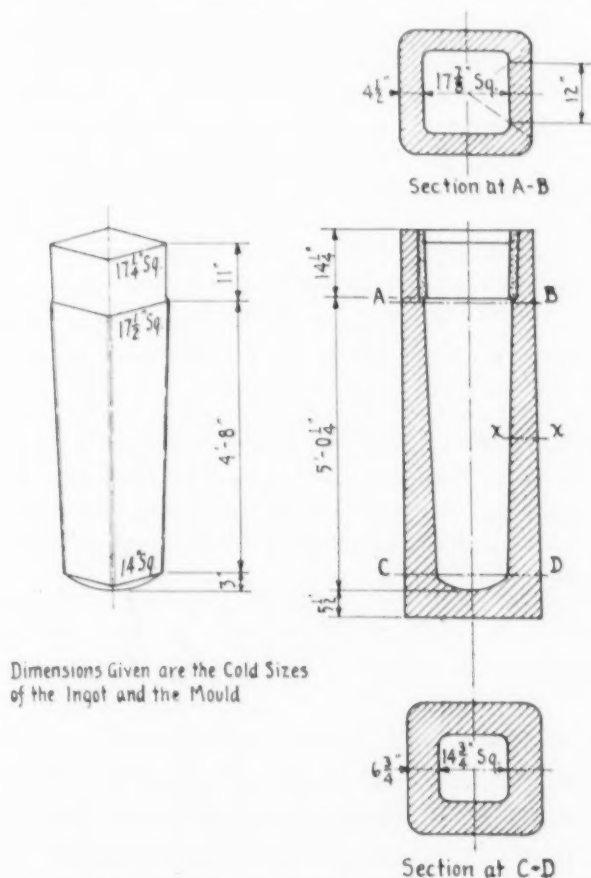


Fig. 14—Diagram of Ingot Mold of Present British Design.

this has now become reasonably stereotyped and examples of present British designs will be found in Figs. 13 and 14. Such matters as the optimum thickness of the wall of the mold and the degree of taper most likely to give the best results, are difficult to fix, since both the type of steel and the layout of the heterogeneity will demand variations in these respects.

(To be continued in May issue of TRANSACTIONS)

THE STABILITY OF METALS AT ELEVATED TEMPERATURES

BY CLAUDE L. CLARK AND ALBERT E. WHITE

Abstract

This research was undertaken for the purpose of answering three questions: first, what relation, if any, exists between the results obtained from short-time tensile and long-time creep tests; second, what are the factors affecting the stability of metals at elevated temperatures; and third, what mathematical relationship exists between the variables encountered in long-time testing.

In regard to the first, it has been concluded that whether or not any relationship exists between these two forms of testing depends entirely upon the temperature range being considered.

In regard to the second, it has been concluded that stability may be increased by increasing the strength of the weakest phase present. That above the equi-cohesive temperature, the amorphous phase, is the weaker, while below, the crystalline phase is the weaker of the two.

In regard to the third, mathematical equations have been developed connecting together stress and time for any particular temperature and any particular metal.

DURING the past few years, considerable time and effort have been expended by many investigators in studying the behavior of metals at elevated temperatures. This interest has been due partly to the demands made by the various industries for information on this subject, and partly to theoretical considerations pertaining to the elastic and plastic nature of metals. Viewed from the practical standpoint, even though operating temperatures have risen to such a point that many plants are now working at temperatures that were unthought of only a few years ago, there is a desire to raise these temperatures still farther. Before this can be safely done, however, more exact information must be

A paper presented before the tenth annual convention of the society, held in Philadelphia, October 8 to 12, 1928. Of the authors, who are members of the society, Dr. Albert E. White is director of engineering research, and Dr. Claude L. Clark is holder of the fellowship in metallurgy established by the Detroit Edison Company, University of Michigan, Ann Arbor, Michigan.

obtained regarding the behavior of metals at elevated temperatures, especially when subjected to continuous stresses.

Two of the outstanding problems at present involved in the use of metals at elevated temperatures are, (1) the development of a short-time test for determining the load-carrying ability of a metal at elevated temperatures, and (2) the determination of the effect of various alloying elements upon the high-temperature stability of alloys.

Considerable work has already been done on the determination of the load-carrying ability of a metal when at elevated temperatures. Up to the present, however, the only sure way of accomplishing this has been by means of long-time creep tests. Since these tests require a considerable amount of time, efforts have been made to relate results obtained from these tests with those obtained from the usual short-time tensile tests. None of these attempts have been very successful. In fact, there is still some question as to whether or not any relationship really exists between the results obtained from these two different kinds of tests.

The second of the two problems has, so far, received but little theoretical attention. It is true that a large number of high-temperature tests, both short-time and creep tests, have been conducted on various alloys and a large mass of data collected, but almost no attempt has been made to explain why certain results have been obtained. Thus investigators have found that additions of chromium increase stability considerably, while nickel has a very slight influence, but no hypothesis has been advanced to account for this phenomenon. It is clear that, if alloys are to be made on a scientific basis for high-temperature duty, it will be necessary to determine the influence of each element upon the stability. Only in this way can the correct combinations of alloying elements be scientifically obtained.

The present investigation was undertaken with a view to furnishing data which might throw light on both of these problems. It was hoped that on the basis of the data obtained for long-time expansion tests, some means might be determined for arriving at a mathematical interpretation for expressing stability of metals in terms of short-time tensile test data. Were these findings subjected to a mathematical interpretation, then the amount of data necessary to obtain the effect of duration, temperature and stress would be materially reduced.

With a view of obtaining the necessary data for such a correlation, short-time tensile and expansion tests were conducted on plain carbon steels of tubular form and on Enduro, an alloy containing approximately 17 per cent chromium and under 0.10 per cent carbon. The carbon steels were chosen, not because it was felt they offered possibilities at the higher temperatures, but because they represent our most common type of steel. Enduro metal was chosen as a typical example of an alloy steel in which the alloying element, insofar as possible, unites with carbon to form complex carbides.

The original results obtained from these tests were then treated mathematically and found to be expressed correctly by means of a formula of the following type:

$$y = ax^b$$

in which y = stress in pounds per square inch, x = duration in hours, a and b = curve constants.

The findings from the expansion tests appeared to suggest that properly determined proportional limit values at temperatures above the equi-cohesive temperature, 842 degrees Fahr. (450 degrees Cent.) for plain carbon steels,¹ and up to at least 1250 degrees Fahr. (677 degrees Cent.),² and at least to 1500 degrees Fahr. (816 degrees Cent.) for Enduro metal,² can be used as a criterion for stability up to a period of 10,000 hours.

A theory is advanced explaining how metals fail at elevated temperatures and why certain alloying elements tend to increase the load-carrying ability of the metals to which they are added. The conclusion is advanced that the two causes of increased stability of metals when above the equi-cohesive temperature are:

1. An increase in the amount of crystalline material as compared to the amorphous material.
2. The formation of compounds which precipitate at both the grain boundaries and within the crystals of the matrix, which, in consequence interferes with the plastic flow.

A brief survey of the literature on long-time tests, indicates that two schools exist:—the one believing that there is a limiting stress below which creep will not continue up to fracture; the other

¹This equi-cohesive temperature is taken as the lowest recrystallization temperature. Jeffries and Archer—*Science of Metals*, pp. 86 and 166.

²The conclusions are limited to these temperature ranges because no experimental work was conducted at higher temperatures.

believing that creep will occur at any stress for a given temperature but at greatly varying rates. When it comes to defining the critical stress at or below which creep will not continue until fracture occurs, the first school is divided into the following three classes:

- 1) those who believe that this limiting-creep value approximates the proportional-limit value as determined in the usual short-time tensile test.
- 2) Those who believe that this limiting-creep value, within a certain temperature range at least, approximates the ultimate-strength value, or some definite fraction of this value, as determined in the usual short-time tensile test.
- 3) Those who have made no effort to relate this limiting-creep value with the values obtained in the usual short-time tensile test.

Considering the methods of expressing test data used by the various investigators in their attempts to determine the existence or non-existence of limiting-creep values, the following systems of plotting are found to be typical:

- 1) Stress versus time.
- 2) Creep versus time.
- 3) Rate of creep versus time.
- 4) Temperature versus time.
- 5) Temperature versus logarithm of time.
- 6) Stress versus logarithm of time.
- 7) Square root of stress versus logarithm of time.
- 8) Stress versus logarithm of the rate of creep.

The application of two or more of the above methods to identical test data often leads to conclusions that are directly opposed to one another. This is not due to the mathematics involved but to the fact that sufficient test data are not obtained to warrant the extensions and extrapolations that are often applied to the curves obtained by the above methods.

METHODS OF PRESENT INVESTIGATION

For the purpose of obtaining data for the determination of mathematical relations between the results from long and short-time tests work was done on two grades of plain carbon steels, one

grade containing 0.13 per cent and the other 0.38 per cent carbon, and on Enduro metal, a high chromium, low-carbon alloy. Their composition was as follows:

SYMBOL	C Per Cent	Mn Per Cent	Si Per Cent	S Per Cent	P Per Cent	Cr Per Cent	Ni Per Cent
A	0.13	0.29	0.20	0.044	0.010		
B	0.38	0.56	0.15	0.028	0.017		
C*	0.37	0.69	0.23	0.020	0.015		
J	0.09	0.34	0.84	0.017	0.021	16.70	0.19

*Bar stock.

Work was also done on various grades of alloy steels, but only for the purpose of throwing light on the factors, which produce stability at elevated temperatures. The table of chemical compositions for the steels used is given in the Appendix.

The two grades of plain carbon steel and Enduro metal were subjected to both short-time tensile and long-time expansion tests. The alloy steels were subjected to short-time tensile tests only, since in this particular phase of the work it was only desired to determine stability through short-time tests.

Procedure for Short-Time Tensile Tests. The short-time tensile test is probably the one most often applied to metals in order to determine certain of their physical properties. The test consists in applying an increasing tensile load to a specimen until rupture of the metal occurs. The total time required for rupture is only a few minutes.

A diagram of the set-up and test specimens used in the short-time tensile tests is shown in Fig. 1. The jaws of the testing machine are replaced by steel blocks containing hemispherical holes into which balls on the end of the specimen holders fit. This arrangement not only eliminates slippage but also insures more perfect alignment than is possible with the usual type of jaw.

The apparatus consists of two circular collars, one fastened to the upper portion of the test piece and the other to the lower part. If it is desired to determine the modulus of elasticity, it is necessary that these collars be fastened at a known distance apart and that the portion of the specimen between them be of uniform cross-sectional area. A pair of vertical rods is fastened to each collar. To the rods leading from the upper collar is fastened the upper aluminum plate which carries the dials, and to the rods leading from the lower collar is fastened the lower aluminum plate. As the load is applied, the plates move apart due to the strain on the



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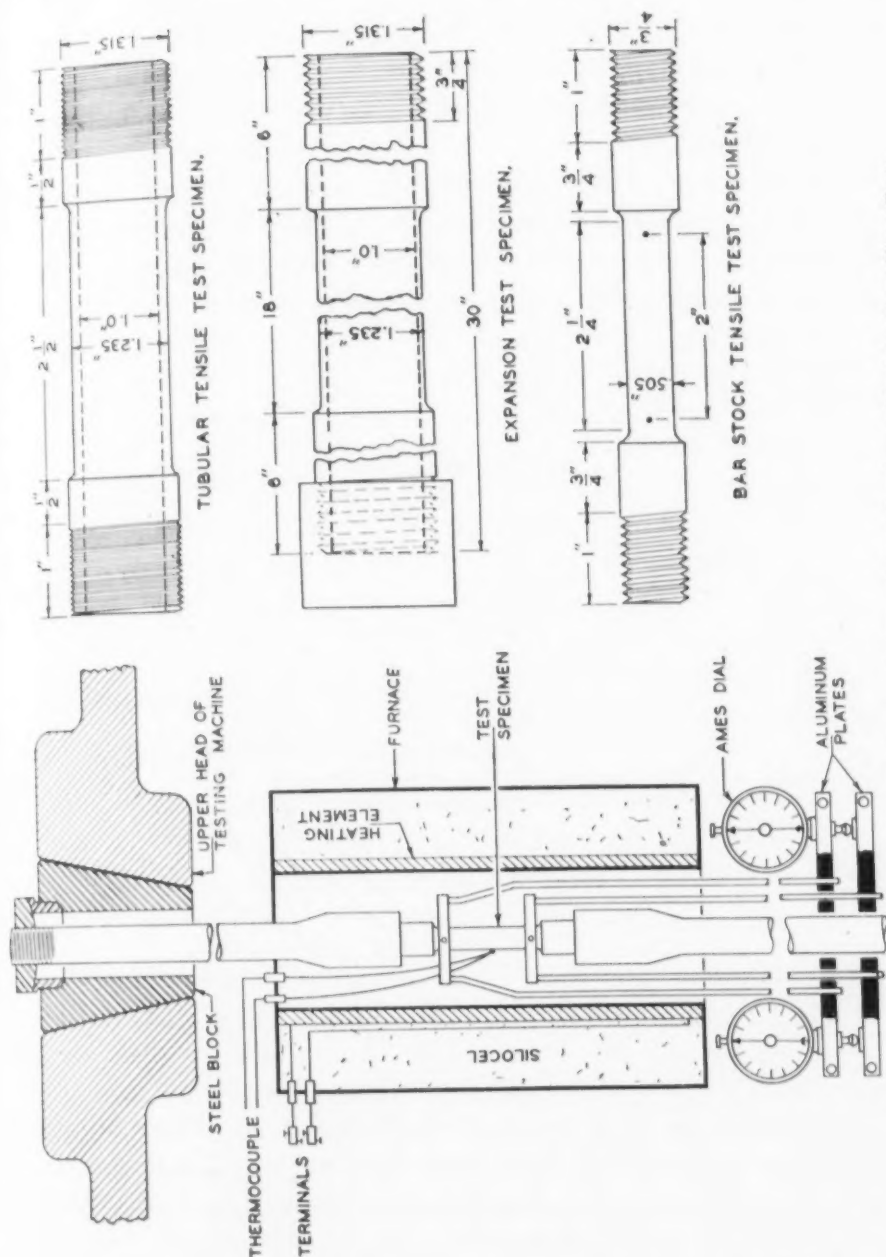


Fig. 1—Diagram of the Testing Machine and Types of Test Specimens Used in Short-Time Tests.

Since the amount of heat loss by radiation and conduction varies over the length of a furnace, it is necessary that the heat be applied non-uniformly if the temperature is to be constant throughout the furnace. This can be accomplished by the irregular spacing of the heating element. The greatest temperature uniformity for vertical furnaces is accomplished by having the windings at the bottom of the furnace the closest together, those at the

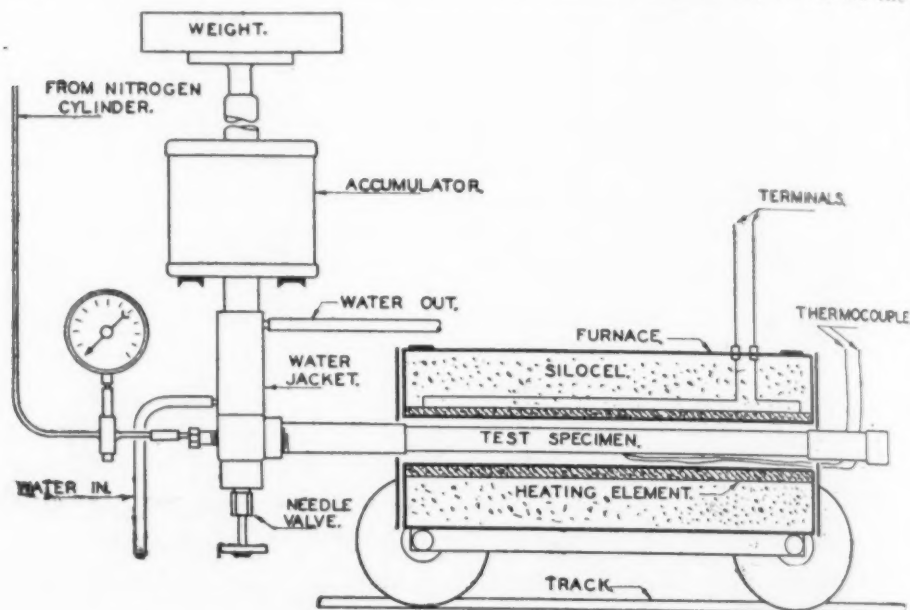


Fig. 2—Diagram Showing Arrangement of Expansion Test Unit.

top farther apart, and those in the center still farther apart. Due to the fact that the temperature distribution throughout the length of the furnace depends upon many factors such as diameter and length of furnace, bottom and top furnace covers, and size of projecting parts, the exact spacing of the winding necessary to give the greatest temperature uniformity can only be obtained by trial. For the first approximation, however, that given by Lynch, Mochel, and McVetty is very useful.

For the horizontal furnaces, the distance between the windings on each end of the furnace was the same and smaller than in the case of the center windings.

Procedure for Expansion Tests. The expansion test was performed by filling a tube with some inert gas, as nitrogen and measuring the pressure with accumulators. The tubes were then brought to the required temperature in electric furnaces and the tempera-

ture and pressure maintained constant. The tests were continued until either rupture occurred or the test was discontinued because it was felt that rupture would not occur during any economic period of operation.

The apparatus used in the expansion test consisted of a pressure accumulator and a seamless-tube test specimen, one end of which was screwed into the accumulator and the other end closed by means of a steel cap. A movable electric furnace was used in heating the tube. A drawing of an expansion test unit is given in Fig. 2.

The pressure accumulator was composed of two pistons of different diameters working against oil, and a reduced section of the larger piston working against nitrogen, the gas used for the production of pressure within the tube.

The type of test specimen used in the expansion test is shown in Fig. 1. It consisted of a seamless tube of the material to be tested 30 inches long with a middle section 18 inches in length slightly reduced. The purpose of this reduction was to facilitate expansion and to eliminate the effects of handling marks produced on the ends of the tubes. It was found that steel caps brazed on to the ends of the tubes gave less trouble from leaks than any other type considered. Cast iron caps became porous in either the brazing or welding operation, and anti-leak preparations as white lead, litharge and glycerine, failed to function properly at the temperatures and pressures employed.

RESULTS OF THE SHORT-TIME TENSILE TEST

The results obtained from the short-time tensile tests conducted on 0.13 per cent and 0.38 per cent carbon steels and on Enduro metal are given in Figs. 3, 4, and 5. The terms in which these results are expressed are defined, according to the International Critical Tables.

An examination of the results presented in Figs. 4 and 5 indicates that, although the 0.38 per cent carbon steel shows the highest proportional limit value at room temperature, the Enduro metal gives superior values at the remaining temperatures considered. The highest tensile strength values were obtained with the 0.38 per cent carbon steel over the entire temperature range.

It will be observed that in the case of the 0.38 per cent carbon

steel as shown in Fig. 4, the test results are given on both tubular and bar stock, which happens to be 0.37 per cent carbon steel. This was done because in an earlier paper by White and Clark the accuracy of the tubular short-time tests was questioned, and tests were therefore made on bar stock for the purpose of bringing out the close agreement which exists between the two shapes of test

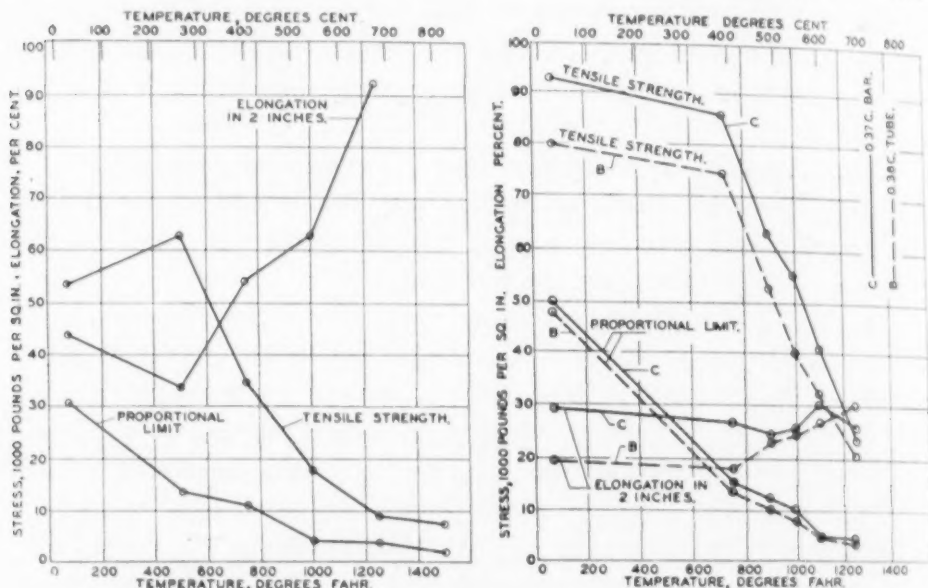


Fig. 3—Properties of 0.13 Per Cent Carbon Steel at Elevated Temperatures as Determined by Short-Time Tensile Tests. Fig. 4—Properties of Carbon Steels at Elevated Temperatures as Determined by Short-Time Tensile Tests.

specimens. It will be noticed that in general all of the values are in close agreement though those recorded for the tubular specimens are slightly below those obtained with the standard tensile test specimen taken from bar stock.

The differences in the values obtained with the two types of test specimens may be due to two causes: (1) wide variations in the thickness of the tube wall, or (2) differences in the temperature distribution. Although, it is felt that the first contributes to the differences obtained, the second is believed to be the most important factor. In the case of tubular specimens, much more uniform temperature conditions can be maintained throughout the entire specimen due to the distribution of the metal over a greater area.

RESULTS FROM THE EXPANSION TESTS

In order to make the results obtained from expansion tests

comparable with those obtained from the short-time tensile and usual long-time creep tests, it is necessary to express them in terms of tensile stress in the tube wall and not as so many pounds internal pressure within the tube. The simplest formula for making this conversion is as follows:

$$\text{Tensile stress} = \frac{\text{pressure} \times \text{internal diameter}}{2 \times \text{wall thickness}} \quad (\text{Equation 4})$$

As pointed out by Wright, however, in his discussion of the former paper by White and Clark, this formula does not take into account either the stresses due to the pressure on the ends of the tube or the elasticity of the material. When these factors are considered, the conversion can best be made by the following formula of Clavarino's:

$$\text{Tensile stress} = \text{pressure} \frac{(4D_1^2 + D_2^2)}{3(D_1^2 - D_2^2)} \quad (\text{Equation 5})$$

in which D_1 = external diameter of reduced section of tube

D_2 = internal diameter of tube.

One of the factors upon which the validity of this formula depends is Poisson's ratio,—the ratio of lateral contraction per unit of diameter to longitudinal extension per unit of length under terminal tension within the elastic limit of the material. It is a well established fact that at room temperatures this ratio is a constant for ordinary steels having a value of $\frac{1}{4}$ to $\frac{1}{3}$. In order to apply the formula to this work, the ratio must also be constant at elevated temperatures. Carrington investigated this ratio at temperatures up to 600 degrees Fahr. (316 degrees Cent.) and found that it stayed within the required limit throughout this range.

It must be realized, however, that the relationship between internal pressure and fiber stress as expressed even by this formula is not rigidly exact. As stated before, the effect of temperature on Poisson's ratio has been investigated only up to 600 degrees Fahr. (316 degrees Cent.) so that the influence of higher temperatures on this ratio is not known. It should also be noted that Poisson's ratio holds only for stresses under the proportional limit of the material in question. In the present work, approximately fifty per cent of the stresses employed were above the proportional limit values. Again, in the derivation of this formula, the material

is assumed to be perfectly elastic, although many investigators are now inclined to consider metals at the higher temperatures as plastic and consider failure to occur by a process of viscous flow. With all of its faults, however, Steward has found this formula to give the best results of any he has considered. It is accordingly

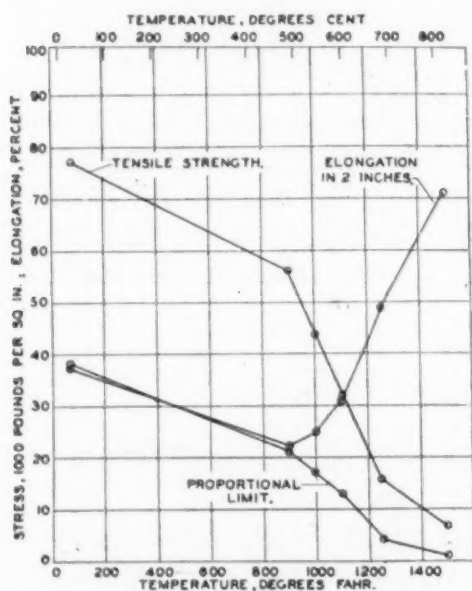
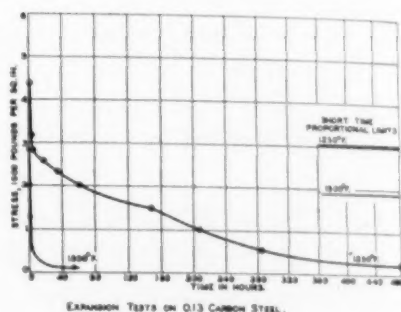


Fig. 5—Properties of Enduro at Elevated Temperatures as Determined by Short-Time Tensile Tests. Fig. 6—Expansion Tests on 0.13 Per Cent Carbon Steel. Curve I at 1250 degrees Fahr. Curve II at 1500 degrees Fahr.



the formula used in the present paper to convert the results of the expansion tests into terms of tensile stress.

For 0.13 Per Cent Carbon Steel. The tests at 1500 degrees Fahr. (816 degrees Cent.) and 1250 degrees Fahr. (677 degrees Cent.) on 0.13 per cent carbon steel were the first expansion tests to be conducted and in the preliminary work the runs were not carried to the point where the tubes failed. At the time these tests were first planned, it was intended to continue the tests only until a marked degree of expansion was obtained and then by means of mathematical interpretations to arrive at conclusions regarding the ability of the metal to withstand expansion at any given temperature for any given pressure during any economic period of operation. It is still believed that this is possible but in order to make the work even more conclusive the tests are now being run until rupture occurs or until it is evident that no rupture will take place during a normal long-time period of operation,

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The results at 1500 degrees Fahr. (816 degrees Cent.), Fig. 6, are of no practical importance because the metal would fail due to oxidation in a relatively short length of time even though it might not fail as a result of creep. For a metal to be serviceable at elevated temperatures it must not only retain appreciable strength at these higher temperatures but must be able to resist oxidation as well. At 1500 degrees Fahr. (816 degrees Cent.), 0.13 per cent carbon steel tubing fulfills neither of the above requirements. The results are nevertheless included in the present

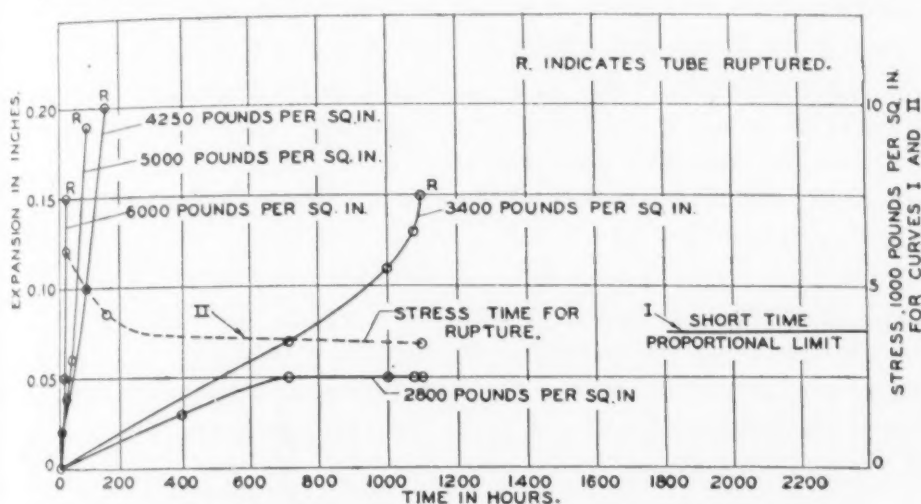


Fig. 7—Expansion Tests on 0.13 Per Cent Carbon Steel at 1100 degrees Fahr.

paper in the hope that they may be of interest. It will be observed that it was necessary to remove the tube subjected to a load as low as 115 pounds per square inch, after 540 hours at heat, because it was so badly oxidized on the surface that it would have failed shortly thereafter.³

The expansion tests at 1250 degrees Fahr. (677 degrees Cent.), Fig. 6 were not continued until fracture of the tubes was obtained. A review of the results, however, shows that with a load below the proportional limit, 2880 pounds per square inch, very marked expansion was obtained in the relatively short period of 18 hours. Comparing this expansion with that obtained in runs in which the

³It should be noted that throughout this report the term "load" implies the tensile stress set up in the tube wall, fiber stress, and not internal pressure to which the tube is subjected. The latter term is the one generally used, but results expressed in that manner depend upon the inside and outside diameters of the tube, while those expressed by the first method are independent of the tube's dimensions. Thus, in the preceding paragraph, 115 pounds per square inch is not the internal pressure to which the tube is subjected but is the tensile stress in the tube wall.

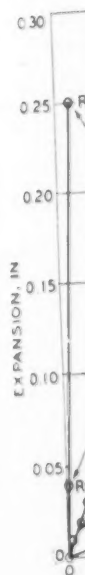
tests were continued until rupture occurred, it is evident that failure would have resulted shortly thereafter.

The results at 1100 degrees Fahr. (593 degrees Cent.) Fig. 7, were obtained by continuing the runs until rupture occurred or until it appeared evident that rupture could not be obtained within a considerable period of time.

As the method used in this case for expressing the results graphically is typical of that employed in all the remaining cases it should perhaps be explained more fully. There are three coördinates used; the horizontal one is time expressed in hours, the vertical one at the left expansion in inches, and the vertical one at the right stress expressed in 1000 pounds per square inch. The solid lines originating at the lower left-hand corner and rising at various rates as they proceed to the right are constant load lines expressing the time rate of expansion of the tubes. The coördinates used for plotting these lines are the vertical one at the left and the horizontal one. The dotted line, designated as the stress-time-for-rupture curve, is obtained by plotting the stress producing rupture against the time required for this rupture to occur. The short, heavy horizontal line at the right is drawn at a stress corresponding to the proportional limit value for this particular metal at the given temperature.

At this temperature, three loads above the proportional limit were used, one equal to the proportional limit, and one below. With the loads above the proportional limit, rupture was obtained in fairly short periods of time, a load of 4250 pounds per square inch, 25 per cent above the proportional limit, producing rupture in 160 hours. With the load equal to the proportional limit rupture was not so readily obtained, 1100 hours being required. With the load below the proportional limit, rupture was not obtained in 1100 hours, and during the last 380 hours no expansion was recorded. This lack of expansion may be due to the method of measuring, as extremely small changes cannot be detected. This last load is but 18 per cent under the proportional limit value as obtained by the short-time tensile test.

From the results obtained at this temperature, a stress-time-for-rupture curve can be drawn as shown in Fig. 7. It is difficult to foretell what may happen to this curve when extended beyond 1100 hours, but, if it be assumed that it becomes asymptotic, it



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would be interesting to make predictions regarding this limiting stress and compare it with the proportional limit value. From the appearance of the curve, it would seem as if a fair value for this asymptotic line would be 2500 pounds per square inch or 26 per cent of the proportional limit value. Later, these results will be treated mathematically in an attempt to arrive at definite conclusions.

Five tests in all were run on the 0.13 per cent carbon steel at 1000 degrees Fahr. (538 degrees Cent.), Fig. 8. In four cases, the load employed was above the proportional limit value obtained in the short-time tensile test. With three of these loads, failure was

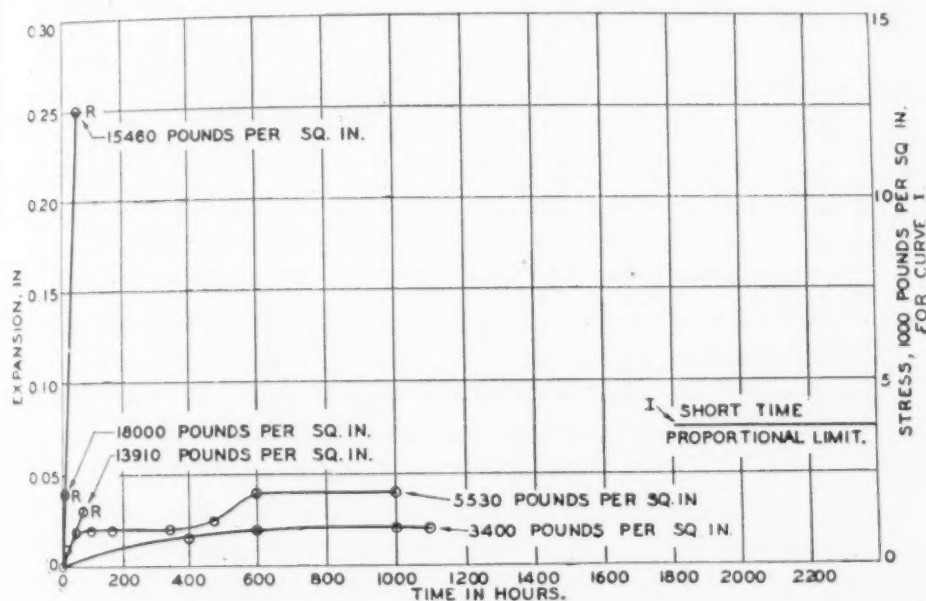


Fig. 8—Expansion Tests on 0.13 Per Cent Carbon Steel at 1000 degrees Fahr.

obtained in relatively short periods of time; while with the load 1000 pounds above the proportional limit value, failure was not obtained after 1000 hours. The values presented do not allow a stress-duration curve to be very accurately drawn, but the limiting stress for this steel at this temperature appears to lie at 3500 pounds or greater.

Attention is called to the fact that two of these tubes failed with an expansion of less than 0.05 of an inch. We offer no explanation of this phenomenon as usually tubes expand at least 0.15 of an inch before they fail.

Only one test is submitted for this steel at 900 degrees Fahr.

(482 degrees Cent.), and this was with a load considerably below the proportional limit value obtained at this temperature. After 1080 hours the expansion was so slight as not to be appreciable by the means employed for measuring expansion.

For 0.38 Per Cent Carbon Steel. One test was conducted with tubing of 0.38 per cent carbon steel at 1500 degrees Fahr. (816 degrees Cent.) but the tube became badly oxidized before rupture occurred. When oxidation occurs, it is difficult to tell whether failure is due mainly to this cause or to expansion. For this reason, further tests at this temperature were not undertaken.

The results of five tests at 1250 degrees Fahr. (677 degrees Cent.) are submitted, Fig. 9. Four of these tests were conducted with loads below the proportional limit as determined by the short-time tensile test on the tube stock. With a load 11 per cent above the proportional limit, 4000 pounds per square inch, failure was obtained in 36 hours; while with a load of 3267 pounds per square inch, 7 per cent below the proportional limit, 405 hours were required to produce rupture. With a load of 1100 pounds per square inch, approximately one-third of the proportional limit value, failure had not yet occurred at the end of 2854 hours but marked expansion was obtained. During the last 854 hours no expansion was recorded, but again this may be due to the fact that the method of measuring expansion is not sufficiently refined and not to the fact that expansion is not continually occurring.

In order to determine the reliability of this method of testing, two tests were undertaken with loads of 2500 pounds and 2200 pounds per square inch. The load of 2500 pounds per square inch required 1090 hours to produce rupture, while the load of 2200 pounds per square inch required 1152 hours. This is considered as a good check on the method employed.

Again if an attempt were made to select a limiting stress from the stress-time-for-rupture curve shown in Fig. 9, it would be in the neighborhood of 1500 pounds per square inch. Although this value is only a deduction, from the mathematical treatment following, it will be shown to be relatively accurate.

The results of three tests at 1100 degrees Fahr. (593 degrees Cent.) are given in Fig. 10. In this case, all the tests were run with loads above the proportional limit as determined on the tubular test specimens. With a load of 10,735 pounds per square



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Fig. 12
Fig. 13

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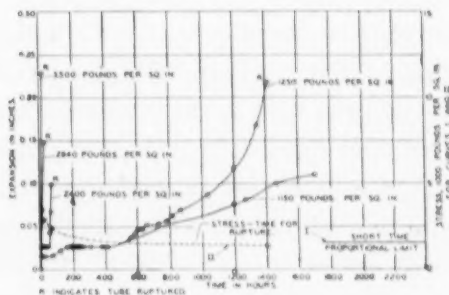
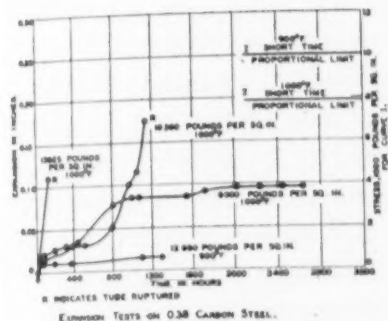
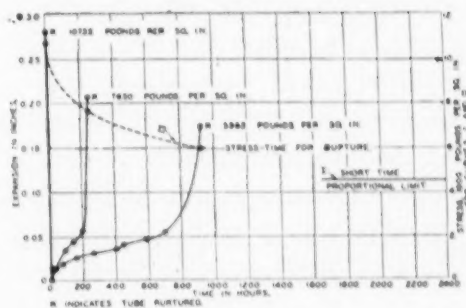
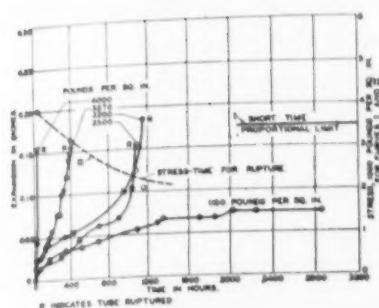


Fig. 9—Expansion Tests on 0.38 Per Cent Carbon Steel at 1250 degrees Fahr. Fig. 10—Expansion Tests on 0.38 Per Cent Carbon Steel at 1100 degrees Fahr.

Fig. 11—Expansion Tests on 0.38 Per Cent Carbon Steel at 900 and 1000 degrees Fahr. Fig. 12—Expansion Tests on Enduro at 1500 degrees Fahr.

inch 138 per cent above the proportional limit, rupture was obtained in 15 hours; while with a load only 32 per cent above the proportional limit, 5960 pounds per square inch, 931 hours were required.

From the stress-time-for-rupture curve it appears that in this case the proportional limit might be the limiting stress below which rupture will not occur, if such a stress does exist.

Three expansion tests were conducted on this metal at 1000 degrees Fahr. (538 degrees Cent.), Fig. 11. In two cases, the loads employed were above the proportional limit values. With both of these loads rupture occurred. With a load of 6000 pounds per square inch, 1875 pounds below the proportional limit, rupture did not occur during the 2636 hours that the test was continued. Though marked expansion occurred in the early part of the test none was recorded during the last 1000 hours.

Sufficient values are not available at this temperature for the construction of a stress-time-for-rupture curve, but from the individual test values it appears that the limiting stress, if such exists, lies in the neighborhood of the proportional limit value.

Only one test was conducted on this tubing at 900 degrees Fahr. (482 degrees Cent.), Fig. 11, and this was with a load of 12,660 pounds per square inch, which is approximately equal to the proportional limit value as obtained from the short-time tensile test at this temperature. Since such a small expansion was obtained even after 1294 hours, it is probable that any limiting stress would be above this proportional limit value.

For Enduro Metal. Five tests were conducted on Enduro metal at 1500 degrees Fahr. (816 degrees Cent.), Fig. 12. Three of these tests were run with loads above the proportional limit and in each case rupture was obtained in a relatively short period of time. With a load of 1250 pounds per square inch, however, only 70 pounds below the proportional limit, 1403.5 hours were required to cause fracture. With a load of 1150 pounds per square inch rupture was not obtained at the end of 1704 hours, but during this period the tube expanded considerably and it is believed would have failed in due time if the test had not been discontinued.

If the time required for rupture be plotted against the stress, the stress-time-for-rupture curve shown in Fig. 12 is obtained. If this curve does become asymptotic to a limiting stress in the higher temperature range, the stress would appear to be about 1000 pounds per square inch. If this is the limiting stress value, then it lies only 420 pounds below the proportional limit or for practical purposes may be considered as the proportional limit.

One of the noteworthy results of the tests at this temperature is the resistance offered to oxidation by these tubes. Even the tube that was exposed to this temperature for 1400 hours showed little sign of oxidation. This is in marked contrast to both the 0.13 per cent and the 0.38 per cent carbon steel tubes, which were both badly oxidized at this temperature.

Five tests were conducted at 1250 degrees Fahr. (677 degrees Cent.), Fig. 13. Three of these were above the proportional limit as determined by the short-time tensile test. With a load of 5970 pounds per square inch, approximately 1000 pounds above the proportional limit, rupture was obtained in 523.5 hours. With a load 100 pounds below the proportional limit, failure occurred in 1147 hours. Again, the results do not show definitely the maximum load that can be carried at this temperature. From the stress-time-for-rupture curve it appears that the limiting stress value,

if such exists, will be around 4000 pounds per square inch. This maximum allowable stress is approximately 800 pounds below the short-time proportional limit.

Considering the results obtained at the above two temperatures, it will be noticed that this metal undergoes a marked increase in strength and load-carrying ability as the temperature is reduced from 1500 degrees Fahr. to 1250 degrees Fahr. (816 to 677 degrees Cent.). The tubes withstood about four times the stress at 1250 degrees Fahr. (677 degrees Cent.) that they did at 1500 degrees Fahr. (816 degrees Cent.). In other words, in decreasing the temperature 250 degrees Fahr. or 20 per cent based on lower temperature and 17 per cent based on upper temperature, the load-carrying ability is increased approximately 400 per cent.

Five tests were conducted at 1100 degrees Fahr. (538 degrees Cent.), Fig. 14, all of which were above the proportional limit as determined in the usual tensile test. With a load of 16,620 pounds per square inch, approximately 3000 pounds above the proportional limit value, rupture was obtained in 1102 hours; while with a load 1000 pounds less, rupture was not obtained in the 2152 hours that the test was continued. From these results, it would appear that any limiting creep value would be in the neighborhood of 15,000 pounds per square inch, about 2,000 pounds above the proportional limit value at this temperature.

No tests were conducted on this metal at temperatures below 1100 degrees Fahr. (593 degrees Cent.) because, due to the limitations in the maximum pressures obtainable, rupture could not be obtained at the lower temperatures.

MATHEMATICAL INTERPRETATION OF RESULTS

It is evident, from the general shape of the stress-time-for-rupture curve, see Fig. 13, that there are two ways in which it may behave as the time is increased. Either the curve may become asymptotic to a given stress line, or else it may continue to slope downwards, continuously approaching the duration axis. If the former condition is true, then there is a limiting stress below which creep will not continue to fracture; if the later condition is correct, then creep continues to fracture for any load considered but at greatly increasing time intervals as the load is decreased.

Although it is practically impossible to differentiate between these two types of curves by visual means the two can be more

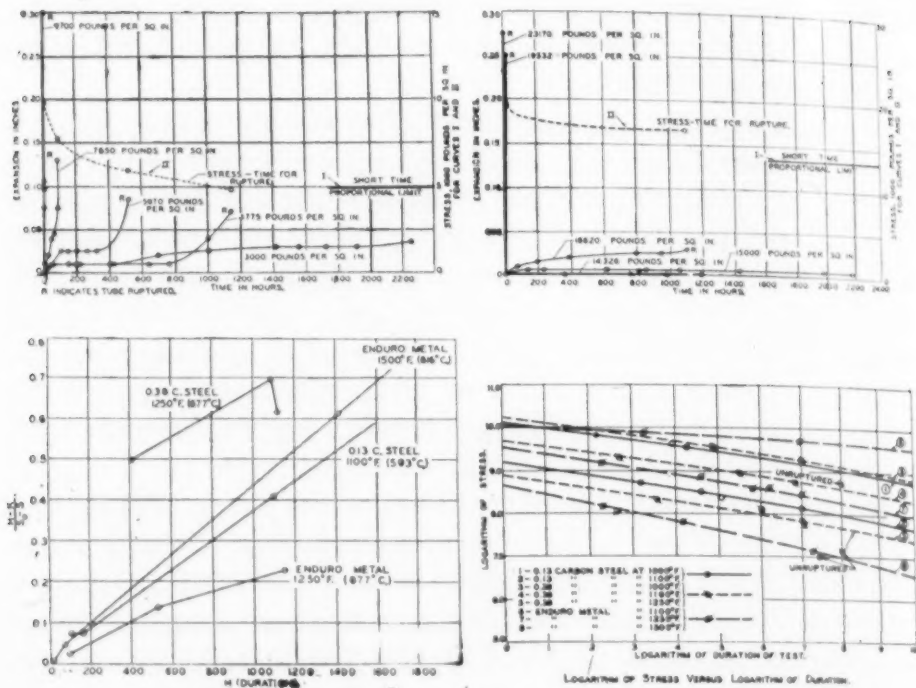


Fig. 13—Expansion Tests on Enduro at 1250 degrees Fahr. Fig. 14—Expansion Tests on Enduro at 1100 degrees Fahr.

Fig. 15—Graphical Method of Testing Hyperbolic Nature of Test Data. Fig. 16—Logarithm of Stress Versus Logarithm of Duration for 0.13, 0.38 and Enduro Steel Tubes at Designated Temperatures.

readily separated from one another by mathematical treatments. Thus, if the curve becomes asymptotic to a given line it can be expressed mathematically as a hyperbolic function, while if it continues to approach the time axis, it is of an entirely different type.

It will now be well to treat the results mathematically in order to determine which of the two conditions the experimental test results most accurately fulfill, the results are mathematically treated in the following paragraph.

Curve Hyperbolic in Nature. Kerr concluded from the shape of the stress-temperature curve, which is similar in form to the stress-duration curve, that it was hyperbolic in nature and so could be expressed by the formula,

$$H - K = (aH + b) (T_0 - T) \quad \text{Equation 6}$$

Changing two of the terms in this expression to suit our needs, we obtain,

$$H - K = (aH + b) (S_0 - S) \quad \text{Equation 7}$$

in which, H = duration corresponding to stress S
 K = duration corresponding to stress S_0
 a and b = curve constants

Transposing the terms in the above equation slightly, we obtain

$$\frac{H - K}{S_0 - S} = aH + b \quad \text{Equation 8}$$

Since a and b are both constants, it is evident from Equation 8 that if H be plotted against $\frac{H - K}{S_0 - S}$ a straight line should be obtained if the above formula correctly represents the test data obtained.

In order to apply the above test, it is necessary to have at least four points at any one temperature. Although in the present research sufficient data are not given in all cases, the formula can be tested on the results obtained on 0.13 per cent carbon steel at 1100 degrees Fahr. (593 degrees Cent.), on 0.38 per cent carbon steel at 1250 degrees Fahr. (677 degrees Cent.) and on Enduro metal at 1500 degrees Fahr. (816 degrees Cent.) and 1250 degrees Fahr. (677 degrees Cent.). The results are shown in Fig. 15.

From Fig. 15, it can be seen that the points obtained by the above method do not fall on a straight line, although in two cases, the agreement is close enough to allow one to state that this formula correctly expresses the results, the disagreement in the other two cases is believed to be sufficiently large to prohibit its use.

Curve Approaches Time Axis. We will now be well to examine the results to determine whether or not the other condition is true, that is, that the "stress-duration curve" tends to approach the time axis. If such is the case, then the result may be expressed mathematically by the equation:

$$y = a x^b \quad \text{Equation 9}$$

in which, y = stress applied
 x = duration of test
 a and b = curve constants

If logarithms be taken of the above equation, the following is obtained:

$$\ln y = \ln a + b \ln x$$

Equation 10

where, \ln = Napierian base logarithm.

From Equation 10, if $\ln y$ be plotted against $\ln x$, a straight line will be obtained whose slope is b and whose intercept on the stress axis (y) is $\ln a$. This simple relationship provides a mean by which the experimental results may be checked to see whether or not Equation 9 expresses them accurately.

In Fig. 16 the logarithms of the results obtained with the various tube materials at the designated temperatures are plotted. Even considering the fact that small differences in logarithm represent relatively large differences in the number themselves, it is believed that the points fall as closely as could be expected on straight lines. Therefore, it appears, as far as these data are concerned, that they are expressed more correctly by Equation 9 than by Equation 7.

From the data given, the equation can be solved for the two constants. From Equation 10, b is the slope of the line and so can be obtained directly from the lines drawn in Fig. 16. The other constant, a , can be obtained from the curves, since $\ln a$ is the intercept on the stress axis, or can be computed from Equation 9 after b is known. The latter method is perhaps the better since intercepts are rather difficult to measure exactly.

Substituting the constant a and b values in the original equation, the following set of equations is obtained for expressing the relationship between stress and duration at constant temperature.

0.13 C. Steel Tubing

$$1100^{\circ}\text{F. (593}^{\circ}\text{C.) : } y = 9566 x^{-0.1477}$$

Equation 11

$$1000^{\circ}\text{F. (538}^{\circ}\text{C.) : } y = 24125 x^{-0.1333}$$

Equation 12

0.38 C. Steel Tubing

$$1250^{\circ}\text{F. (677}^{\circ}\text{C.) : } y = 7670 x^{-0.1600}$$

Equation 13

$$1100^{\circ}\text{F. (593}^{\circ}\text{C.) : } y = 16300 x^{-0.1394}$$

Equation 15

$$1000^{\circ}\text{F. (538}^{\circ}\text{C.) : } y = 27275 x^{-0.1375}$$

Equation 14

Enduro Metal Tubing

$$1500^{\circ}\text{F. (816}^{\circ}\text{C.) : } y = 5560 x^{-0.2061}$$

Equation 16

$$1250^{\circ}\text{F. (677}^{\circ}\text{C.) : } y = 14400 x^{-0.1363}$$

Equation 17

$$1100^{\circ}\text{F. (593}^{\circ}\text{C.) : } y = 26545 x^{-0.0483}$$

Equation 18

Two characteristics of the curves in the logarithmic charts should be noted: (1) all the lines, with the exception of those expressing the results obtained on Enduro tubing at 1500 degrees Fahr. (816 degrees Cent.), and 1100 degrees Fahr. (593 degrees

Cent.) appear to be parallel or very nearly so; (2) the intercepts on the stress axis are directly proportional to the temperature difference of the lines. Due to the limited amount of data presented, these characteristics can as yet be considered only as tendencies and not as established facts.

The question now arises as to whether or not parallel lines would be obtained throughout the entire temperature range. French's results on creep tests at atmospheric temperature show the existence of a sharp division between stresses which will and will not cause failure at this temperature. For example, on his tests with 0.24 per cent carbon steel at 70 degrees Fahr. a stress of 54,000 pounds per square inch produced fracture almost instantaneously, while a stress of 53,400 pounds failed to produce failure in the 1000 hours during which the test was conducted. If these results should be plotted logarithmically, the resulting line would be practically horizontal, having only a very slight negative slope.

As a result of this test it may be concluded that logarithmic plotting of creep results at atmospheric temperature gives practically horizontal lines, and that as the temperature is raised the slope of the lines is increased negatively. Considering the small change in slope in increasing the temperature from atmospheric to 1000 degrees Fahr. (538 degrees Cent.), it appears that the change of slope with temperature is extremely small and so lines covering only a relatively small temperature range would appear to be parallel.

The fact that the intercept difference of these lines appear to be proportional to their temperature difference should also be considered throughout the entire temperature range. If this relationship holds throughout the entire temperature range, then from the results available it will be possible to compute the load required to produce rupture in practically zero time at atmospheric temperature. On this basis the following values are obtained for the three steels considered:

0.13 per cent carbon steel—	200,000,000 pounds per square inch
0.38 per cent carbon steel—	4,000,000 pounds per square inch
Enduro Metal—	900,000 pounds per square inch

Since the differences between these values are so much greater than what one would expect from a comparison on their other properties, it appears that the proportionality between intercept

differences and temperature difference does not continue to exist down to atmospheric temperature.

There is also a possibility that the above two characteristics of the logarithmic lines may hold only over certain temperature ranges. It is now a well established fact that metals behave differently in many respects above a certain temperature, known as the equi-cohesive temperature, from what they do below it. It may be that this relationship only holds at temperatures above this equi-cohesive temperature, but as no work has been done at temperatures below, this remains an open field for future investigation.

Comparison of Creep Stresses. Comparisons between the stress-duration curve in its extrapolated portions with values obtained from the mathematical formula, above developed, in the same section of the curve are shown in Figs. 17, 18 and 19. For comparison, the respective proportional limit values are also included.

From these Figs. it is seen that the stress required to produce rupture in 100,000 hours is below the corresponding short-time proportional limit value in every case, except two, that of 0.13 per cent carbon steel at 1000 degrees Fahr. (538 degrees Cent.) and that of Enduro metal at 1100 degrees Fahr. (593 degrees Cent.). Considering the stress required to produce rupture in 10,000 hours, two are substantially equal to the proportional limit value, two above, and the remaining below. The computed values all compare very favorably with the imaginary limiting stresses taken from the stress-duration curve. The limiting stress values taken from the stress-duration curve correspond very closely to the proportional limit values in all cases except two.

Although the findings in 0.13 per cent carbon steel at 1000 degrees Fahr. (538 degrees Cent.) and on Enduro metal at 1100 degrees Fahr. (593 degrees Cent.) are included, it is believed that these metals under the given conditions of structure may be in the equi-cohesive temperature zone, in which case the formula developed does not hold, as this formula is for temperatures above and not below the equi-cohesive temperature. It is well recognized that for ordinary steels there is a sharp break in the physical properties at temperatures in the neighborhood of 700 degrees Fahr. (371 degrees Cent.) and this break is without doubt due to the properties existing above and below the equi-cohesive temperature. Therefore, although these values are incorporated it is believed that

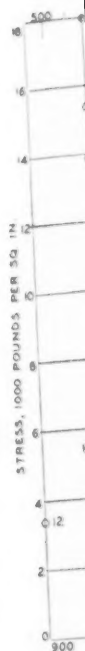
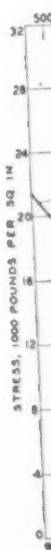


Fig. 17
0.13 Per Cent Carbon Steel at 1000 degrees Fahr.



Enduro Metal at 1100 degrees Fahr.

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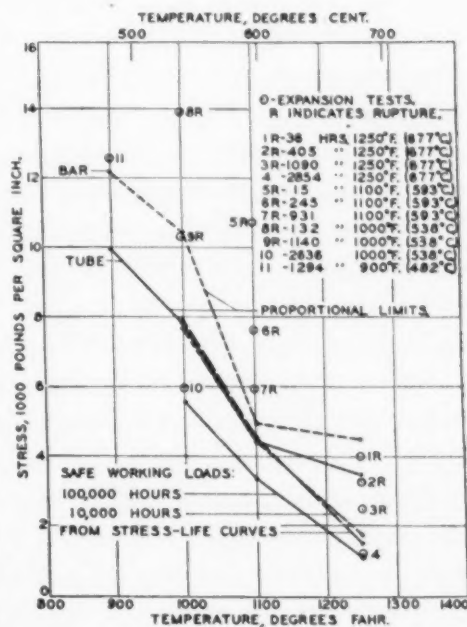
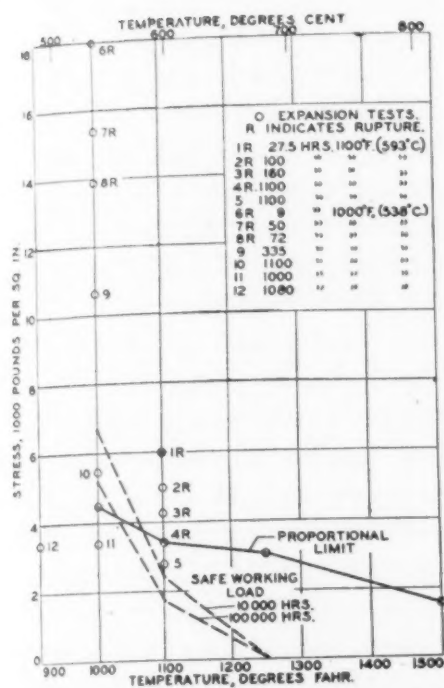


Fig. 17—Expansion Results, Safe Working Loads, and Proportional Limit Values, for 0.13 Per Cent Carbon Steel Tubing. Fig. 18—Expansion Results, Safe Working Loads, and Proportional Limit Values for 0.38 Per Cent Carbon Steel Tubing.

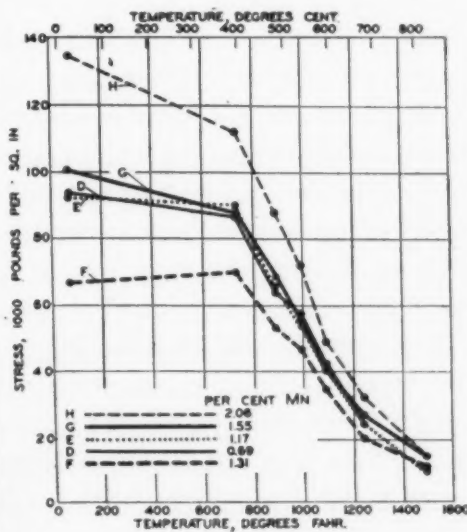
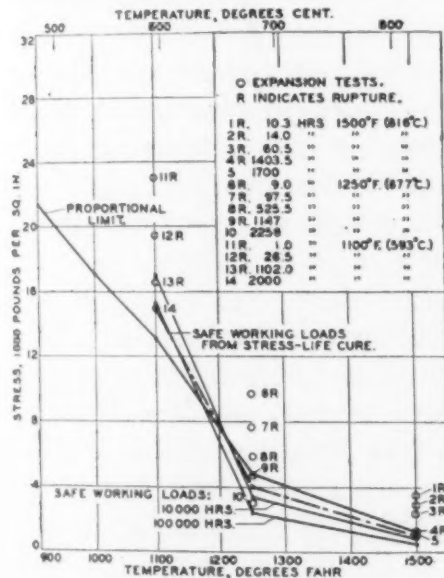


Fig. 19—Expansion Results, Safe Working Loads, and Proportional Limit Values for Enduro Steel Tubing. Fig. 20—Tensile Strength Curves of Various Carbon-Manganese Steels at Elevated Temperatures.

for the purpose of the mathematical interpolation they should be disregarded.

It should also be remembered that the values representing the loads required to produce rupture in 10,000 and 100,000 hours were computed and not obtained experimentally. In obtaining the equations necessary for the computations, it was necessary to assume that the same mathematical relationship which held for the stress-duration-curve through the range of experimental testing, continued to hold as the time was indefinitely increased. Although the authors believe this to be the case, they recognize the propriety of additional experimental work before a too dogmatic acceptance of the results can be expected.

The comparison of proportional limit and safe working load values would lead one to conclude that even though the proportional limit value is not a rigid dividing line above which creep proceeds to fracture and below which it does not, it approximates the stress at temperatures above the equi-cohesive range below which failure is not to be expected within a period of 10,000 hours or so.

Physical Meaning of Mathematical Results. Hooke's Law states, that, if a body is distorted within its limits of elasticity, then the stress producing this distortion is proportional to the deformation produced. Thus, if a stress of 1000 pounds applied to a perfectly elastic body produces a strain of 0.001 inch, then as long as the stress is held constant, the strain will also remain the same provided the given stress is not above the proportional limit.

Solely on the basis of definition, the true proportional limit value for a perfectly elastic body must be the limiting stress below which perfect proportionality of stress and strain exists. If the word creep be assumed to represent a more or less slow deformation continuing to rupture, it follows from this definition that for such bodies creep cannot occur below the elastic limit. It has often been found that metals can withstand stresses above the proportional limit value for extended periods of time. This occurs only at relatively low temperatures, however, and may be due to the fact that the initial deformation produces strain hardening, so that the proportional limit is raised until it has a value above that of the applied stress.

In the expansion tests described in this paper, it has been found experimentally that creep proceeded to fracture in some cases with stresses below the proportional limit as determined by the usual short-time tensile test. This phenomenon can only be accounted

for on one of the following assumptions: (1) the proportional limit as usually determined by the short-time tensile test is not the true proportional limit; (2) metals at the elevated temperatures considered are not perfectly elastic bodies and so do not possess true proportional limits. Although these two cases may appear identical, they are entirely different, for if the first is true, then failure will not occur below a certain definite stress, while if the second is correct, then failure will occur with any load if sufficient time is allowed.

From the mathematical interpretation of the test results obtained by means of the expansion tests, it was found that creep would continue to fracture for any load at a given temperature above the equi-cohesive temperature but at greatly different rates depending upon the temperature and relative magnitude of load. From this, it would appear that the metals under consideration at temperatures of 1000 degrees Fahr. (538 degrees Cent.) or above do not possess true proportional limits, and so are not perfectly elastic bodies. What the case is at temperatures below 1000 degrees Fahr. (538 degrees Cent.) is a matter of conjecture. In other words, in the above classification, the continuous creep at certain elevated temperatures must be explained by the second case and not by the first.

In order to account for this imperfect elasticity of these metals at certain elevated temperatures, it will be well to examine their inner structure. It is now generally accepted that metals consist of crystals, each of which is completely surrounded by a layer of amorphous material; also that at ordinary temperatures this amorphous material is much harder and stronger than the crystalline material. As the temperature is raised, however, the amorphous material weakens at a much more rapid rate than the crystalline material, so that at a certain temperature the boundaries are weaker than the crystals themselves.

Applying these facts to tensile testing, practically 100 per cent of the deformation of ordinary steels at room temperature takes place within the crystal and fracture takes place through the crystals and not in the crystal boundaries. Since crystals are perfectly elastic bodies, a definite proportional limit can be obtained at atmospheric temperature and this value, if correctly determined, will be independent of the rate of application of load.

As the temperature of testing is increased, however, the conditions are entirely changed. A temperature is finally reached where the amorphous boundaries become more plastic and considerable flow takes place in this amorphous material. Time is required, however, in order to allow this flow to occur, and in short-time testing the load is applied so rapidly that the amorphous material is able to resist it for the necessary time and a large fraction of the deformation occurs within the crystal, which causes a proportional limit to appear to exist. If more time is allowed, flow does occur in the boundary material and the metal will fail due to viscous flow in the amorphous material rather than by a process of slip within the crystal itself.

The question now arises as to what temperature is required to produce a viscous flow. From the results of the expansion tests here presented, it is evident that viscous flow was obtained in most cases. Also, the mathematical formulae, which were found to express the test results, are equations of viscous flow and not of slippage which halts at a definite stress. From this it can be concluded that at temperatures of at least 1000 degrees Fahr. (538 degrees Cent.) or above, the metals considered fail by a process of viscous flow of the material in the grain boundaries rather than by slippage within the crystals themselves.

EXPLANATION OF FACTORS AFFECTING STABILITY AT ELEVATED TEMPERATURES

The question now arises as to why Enduro metal showed superior load carrying ability to either of the plain carbon steels, and why the 0.38 per cent carbon steel was superior in this respect to the 0.13 per cent carbon steel. On the basis of the short-time tensile properties reported in the appendix, the effect of various other alloying elements on load carrying stability will also be considered.

In order to predict the influence of alloying elements upon the load carrying ability of metals when at elevated temperatures, it is necessary to know what effect these added elements exert upon the existing structure of the metal. Also in order to be better able to predict which structures impart the greatest load carrying power, it is necessary to understand how increased temperatures affect the manner in which metals deform and fracture.

It is a well established fact that at atmospheric temperatures fracture of ordinary steels proceeds through the grains and not around them. This is due to the greater strength of the amorphous boundaries. Thus any factor tending to decrease grain size, and thereby increase the amount of amorphous material present, will increase the strength of the steel when at room temperature. Likewise, any factor tending to increase the strength of the crystals themselves will increase the strength of the steel when at room temperatures.

As the temperature of testing is increased, however, the amorphous material weakens at a more rapid rate than that of the crystal proper, and a temperature is finally reached where, for these metals, fracture proceeds around the grains and not through them. This temperature Jeffries calls the equi-cohesive temperature. In general, this corresponds to the lowest recrystallization temperature of the alloy when in a severely cold worked state. Certain factors change its location, probably the most important of which is the rate of application of load.

In regard to the dependence of this temperature upon the rate of application of the load, Jeffries reports the equi-cohesive temperature of copper with loads producing rupture in 3 seconds as 1742 degrees Fahr. (950 degrees Cent.); while for very slow loading it is in the neighborhood of 572 degrees Fahr. (300 degrees Cent.).

French in discussing the fractures of plain carbon steels under varying rates of application of load at various temperatures (300 to 865 degrees Fahr.) (149 to 463 degrees Cent.) states, "Minor differences appear under certain conditions, such as, for example, the apparent tendency of the fracture to follow more deeply along the grain boundaries and particularly at the junctions of ferrite and pearlite when the steel is broken under slowly increasing stress." Even though the length of time employed in the above tests was hardly sufficient to classify them as long-time creep tests, at least as they are now known, yet this tendency of fracture to follow the grain boundaries was observed.

In this same connection, Wood states: "We lose strength at elevated temperatures because due to inadequate anchorage of the crystals we fail to get the most work out of the crystals themselves." This same author reports 600 degrees Fahr. (316 degrees

Cent.) as being the temperature at which noticeable weakening of the grain boundaries of plain carbon steels occurs under prolonged stress.

The cases cited above are sufficient to show, that, before the effect of alloying elements upon the stability of metals can be determined, it will be necessary to know over what temperature range and for what length of time this stability is desired, i. e., those elements the addition of which may cause increased stability up to temperatures of 600 degrees Fahr. (316 degrees Cent.) or so, are not necessarily the ones that will have the greatest effect in increasing the load-carrying ability at the higher temperatures.

If the stability is desired below the equi-cohesive temperature, then the following two ways are open:

(1) The addition or formation of any element, compound, or constituent which does not enter into solid solution with the matrix, but which by the very nature of its presence interferes with the crystal slippage will tend to increase the stability at elevated temperatures of the material of which it forms a part.

(2) The addition or formation of any element, compound, or constituent which tends to increase the amount of amorphous material existing between the grains of crystals in metals and like substances tends to increase the stability of the materials of which the amorphous substance forms a part at all temperatures below those of the equi-cohesive range.

If the stability is desired at temperatures much above 600 degrees Fahr. (316 degrees Cent.) for plain carbon steels, then other factors must be taken into account. Due to this temperature range, which is in the vicinity or slightly above the equi-cohesive range, the amorphous phase becomes the weaker and deformation occurs in the grain boundaries. For increasing the stability above the equi-cohesive temperature, which for steel is 600 degrees Fahr. (316 degrees Cent.) or higher, the following two ways are open:

(1) The addition or formation of any element, compound, or constituent which decreases the relative amount of amorphous material will increase the load carrying ability.

(2) The addition or formation of any element, compound, or constituent which strengthens the amorphous boundaries by interfering with plastic flow will increase the load carrying ability.

The ratio of amorphous to crystalline material in any crystal-

line body is determined by the grain size. Thus a cast metal which usually has a greater grain size than a forged or heat treated metal should tend to show greater load-carrying ability at temperatures above the equi-cohesive temperature.

There are some results available to support this statement. For example, in the work reported by Malcolm in which results were reported on both cast and rolled metal of approximately the

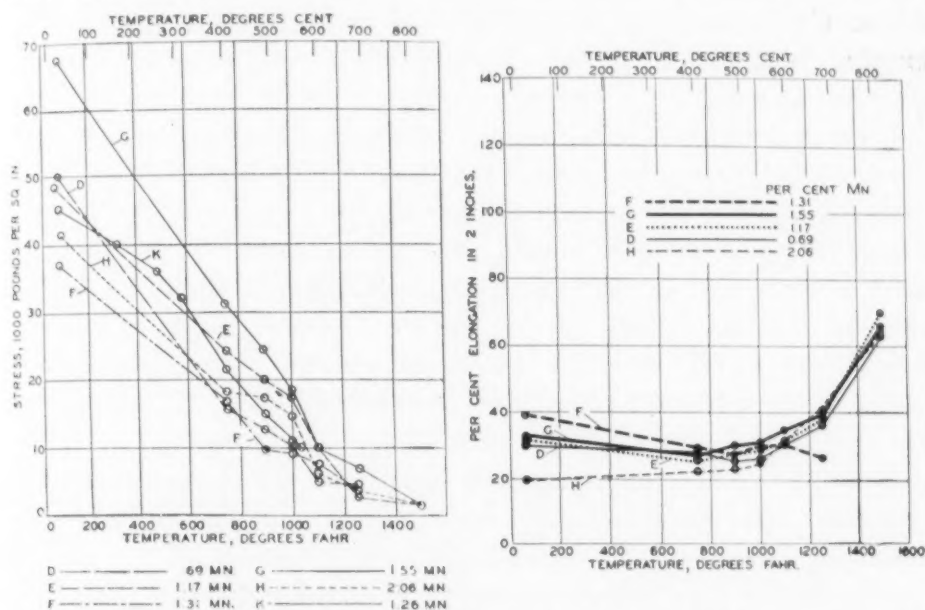


Fig. 21—Proportional Limit Values of Manganese Steels at Elevated Temperatures. Fig. 22—Elongation Curves of Various Manganese Steels at Elevated Temperatures.

same composition, the material in the cast condition was able to withstand a stress of 13,000 pounds per square inch for 2400 hours before rupture occurred; while in the case of the rolled metal, a stress of only 8750 pounds per square inch produced failure in slightly less than 1500 hours. These tests were conducted at 1000 degrees Fahr. (538 degrees Cent.) on chromium-nickel steel.

On the basis of the hypotheses stated above setting forth ways and means of increasing the strength of the material in the amorphous state, it would appear that there should be two ways of increasing the strength of the amorphous boundaries,—either by the addition of a substance which dissolves in the layer and gives it greater strength, or by the addition of a substance which forms compounds that precipitate in this layer and exert a keying action in it against flow. On the basis of the hypothesis stated above

setting forth ways and means of increasing the strength of the material in the amorphous boundaries, it has been observed that carbides fulfill these conditions and on the basis of this assumption it is possible to explain why Enduro metal exhibited a load carrying ability superior to that of both of the plain carbon steels investigated, and why the 0.38 per cent carbon steel was superior to the 0.13 per cent carbon steel. They would also enable one to foretell which elements or combination of elements should produce the greatest increase in the load-carrying ability of the metal to which they are added.

Plain Carbon Steels. It will perhaps be best to consider first plain carbon steels in regard to the effect of structure on load-carrying ability at elevated temperatures. In the lower carbon steels, the 0.13 per cent carbon steel for example, the ferrite is in such a large excess that it forms a complete network about the pearlite crystals. These steels will thus exhibit the properties of ferrite, and as this material is soft and relatively weak, such steels could not be expected to exhibit marked load-carrying ability.

In the case of the 0.38 per cent carbon steel, the final structure of a slowly cooled specimen will consist of the same two metallographic constituents, ferrite and pearlite, but in entirely different proportions. While 0.13 per cent carbon steel consists of approximately 84 parts of ferrite to 16 parts of pearlite, 0.38 carbon steel consists of about 55 parts of ferrite to 45 parts of pearlite. In other words, the proportion of pearlite has been increased approximately three times. Therefore in this case there is not sufficient excess ferrite to surround the pearlite completely, so that the steel displays the properties of pearlite plus ferrite, rather than those of ferrite alone. On this basis the 0.38 per cent carbon steel might be expected to exhibit the superior load-carrying powers at all temperatures below the thermal critical range. Such has been found to be the case, as is shown in Figs. 9, 10 and 11.

Chromium Steels. Since Enduro metal is a high-chromium steel, it will be well to consider chromium steels in general before discussing this particular one. Chromium dissolves in iron in all percentages forming solid solutions with the ferrite. In the presence of carbon, however, it forms double carbides with the iron.

It is a well-established fact that the first tendency of chromium is to combine with carbon to form carbides and that no chromium

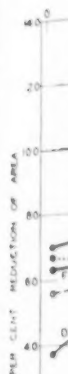


Fig. 9. Temperature

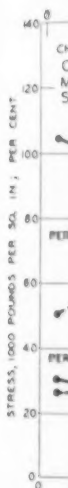


Fig. 10. Properties

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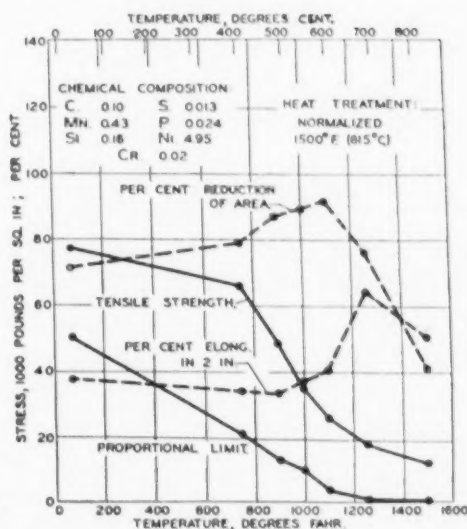
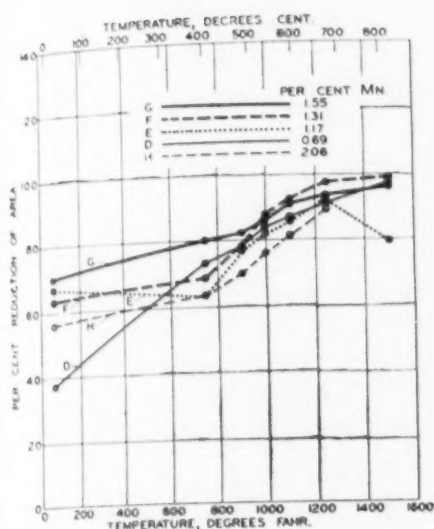


Fig. 23—Reduction of Area Curves of Various Carbon-Manganese Steels at Elevated Temperatures. Fig. 24—Tensile Properties of a Nickel Steel at Elevated Temperatures.

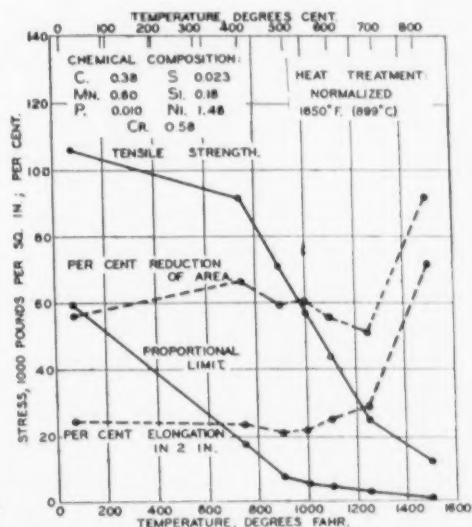
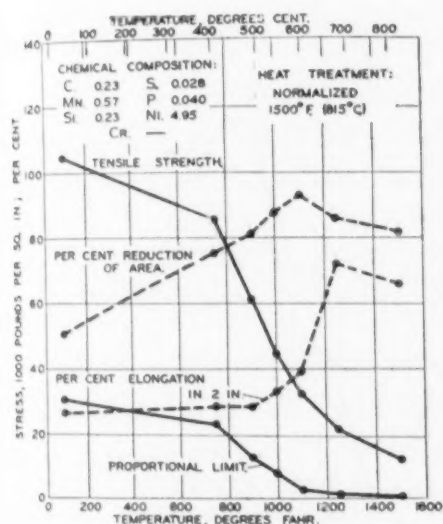


Fig. 25—Tensile Properties of a Nickel Steel at Elevated Temperatures. Fig. 26—Tensile Properties of Chromium-Nickel Steel at Elevated Temperatures.

forms a free solid solution with iron unless insufficient carbon exists for the formation of the carbides. The carbides which are deposited within the crystal will interfere with the process of crystalline slippage and for that reason materially increase the stability of the alloy at all temperatures below the equi-cohesive temperature. The carbides which are precipitated at the grain boundaries will interfere with the plastic flow of this material and so increase the load-carrying ability of the metal above the equi-cohesive tem-

perature. Therefore, steels which contain chromium in sufficient amounts manifest a much greater stability at all temperatures than plain carbon steels, a statement which is supported by the findings shown in Figs. 12, 13 and 14.

Likewise, treatments which produce the free carbides show greater strength in the alloy in which they are found than treatments which throw the carbides into solid solution, as evidenced by the superior results of 8 per cent chromium, 8 per cent tungsten steel, Fig. 30, the constituents of which in the annealed state are ferrite and pearlite plus carbides, over and above those of the two alloys, one containing 20 per cent of chromium and 8 per cent of nickel, Fig. 28, and the other 8 per cent of chromium and 20 per cent of nickel, Fig. 29, both of which because of their compositions exist in the austenitic condition even in the annealed state. To be sure, in the first steel 8 per cent tungsten is present, and in the other two relatively large percentages of nickel, but the structure of the first steel shows free carbides while the last two are both austenitic,—solid-solution alloys.

Although experimental data obtained from long-time expansion tests are not presented on any other steels except those already mentioned, it should be possible on the basis of the assumptions made to predict regarding the load-carrying ability of various alloy steels. The alloying elements to be considered are nickel, tungsten, manganese, molybdenum, vanadium, and various combinations of these elements.

Nickel Steels. Nickel dissolves in steel at atmospheric temperatures, especially in the ferrite crystals. Nickel additions produce a strong tendency towards the formation of austenitic alloys. Therefore slowly cooled nickel steels can have various structures, as pearlitic, martensitic, or austenitic, depending on the relative amounts of nickel and carbon present, and indirectly at least within certain ranges, depending on the heat treatment.

With low-carbon nickel steels in the pearlitic zone, one of the effects of the addition of nickel is to reduce grain size. At atmospheric and slightly elevated temperatures, this reduced grain size improves the strength of the alloy due to the presence of larger amounts of amorphous material. At elevated temperatures above the equi-cohesive temperature this reduced grain size results in a weakening of the alloy if the rate of application of the

load is very small, due to increasing the amount of the amorphous phase, which is the weakest.

Considerable percentages of nickel will produce an alloy with a structure entirely austenitic. Since austenitic steels exhibit low proportional limit values even at room temperatures, it can hardly be expected that they will show unusually high values at elevated temperatures. For this reason, they are seldom used where considerable stress is to be withstood at medium elevated temperatures, as between 1000 degrees Fahr. (538 degrees Cent.) and 1500 degrees Fahr. (816 degrees Cent.).

Dickenson has reported the results of long-time creep tests on a carbon and nickel-chromium steel both containing the same percentages of carbon. Moreover, the nickel-chromium steel, 3.63 per cent nickel, 0.57 per cent chromium, contains so little chromium in comparison to the nickel that this steel may be considered as a nickel steel. On the basis of what has been said, this nickel-chromium steel should prove inferior to the straight carbon steel with respect to its load-carrying ability at high temperatures above the equi-cohesive range. This difference in stability values would be expected to increase as the time of testing was increased. An examination of the results reported by Dickenson show clearly that such is the case.

From this, it can be concluded that plain nickel steels do not offer possibilities of having high load-carrying ability at elevated temperatures above the equi-cohesive range. The results of short-time tensile tests at elevated temperatures on two nickel steels differing only in carbon content are given in the appendix and Figs. 24 and 25.

Chromium-Nickel Steels. The addition of both chromium and nickel to a steel should give an alloy having properties which are a combination of those mentioned in the above two cases. Nickel, dissolving mainly in the iron, will increase the resistance to oxidation of the alloy while the chromium, being a carbide former, will tend to increase the load-carrying ability of the alloy.

Malcolm has published the results of long-time creep tests on chromium-nickel alloys in both the cast and rolled conditions. The metal in the rolled condition shows load-carrying ability nearly equal to that of the 0.38 per cent carbon steel, a steel containing 0.08 per cent more carbon than the chromium-nickel steel, reported

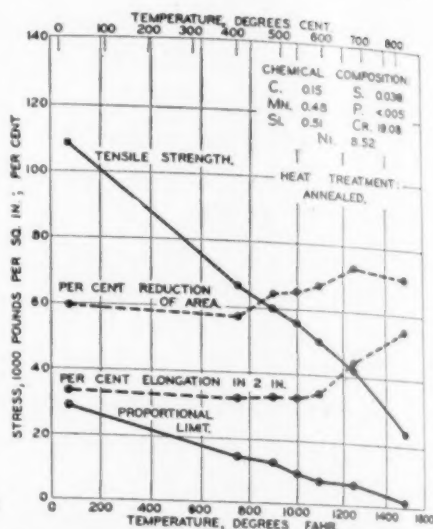
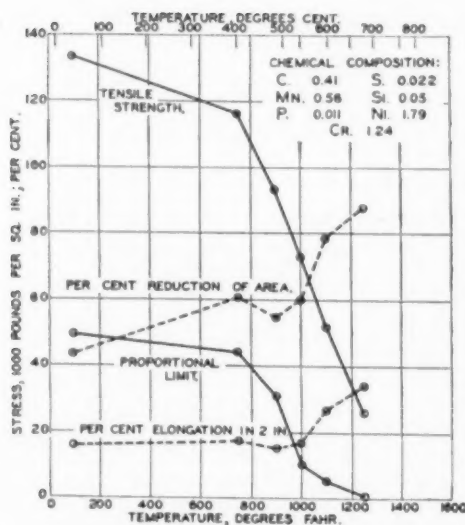


Fig. 27—Tensile Properties of Chromium-Nickel Steel at Elevated Temperatures. Fig. 28—Tensile Properties of Chromium-Nickel Steel at Elevated Temperatures.

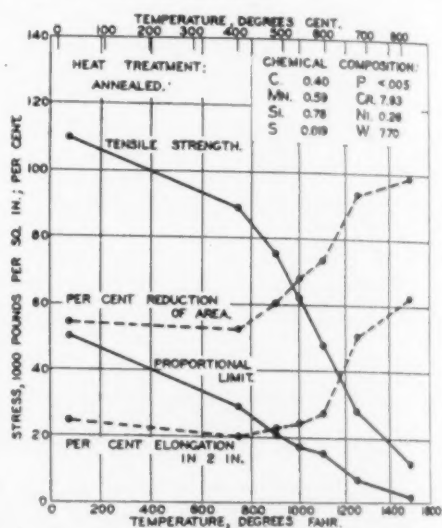
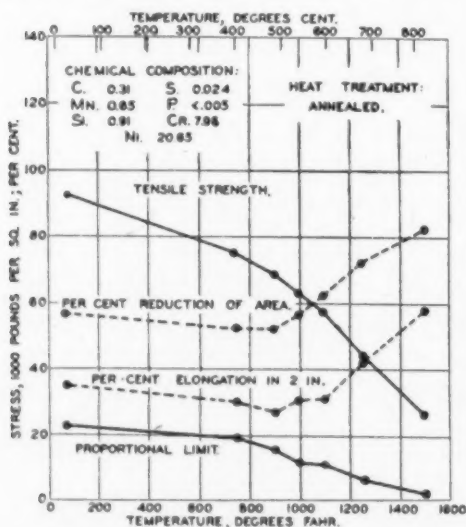


Fig. 29—Tensile Properties of Nickel-Chromium Steel at Elevated Temperatures. Fig. 30—Tensile Properties of Chromium-Tungsten Steel at Elevated Temperatures.

in this paper. In the cast condition, however, the strength exhibited by this alloy steel in the creep test, was much superior to that shown by the straight carbon steel. It must be remembered that in these alloys both the nickel and chromium contents were relatively low.

In Figs. 26 and 27 and the appendix the high-temperature properties, as determined by short-time tensile tests of two chromium-nickel steels, are given. These steels contain practically the same percentages of both carbon and nickel, but the chromium con-

tent of one is approximately twice that of the other. From what has been said, the alloy containing the higher percentage of chromium would be expected to exhibit the higher proportional limit and tensile strength values. An examination of the figures will show that such is the case.

Tungsten Steels. Tungsten combines with the carbon rather than with the iron when added to steel. Since it is similar in this respect to chromium, the tungsten steels might be expected to have load-carrying properties somewhat similar to those shown by the chromium steels.

The effect of tungsten upon the properties of metals at high temperatures is shown in Fig. 30. It is true that this particular alloy also contains about 8 per cent chromium, but if these results be compared to those from Silcrome, an alloy also containing from 8 to 9 per cent chromium, then the effect of tungsten can be determined. It is true that Silcrome contains from 3 to 4 per cent silicon, but this element being similar in its behavior to nickel does not affect the load-carrying properties to any great degree. A comparison of the results obtained with these two steels will show the tungsten bearing steel to have the superior properties, especially at the higher temperatures.

Manganese Steels. Manganese is also similar to chromium in that it combines with the carbon rather than with the ferrite. For this reason, the addition of this element should also increase the load-carrying ability of metals at elevated temperatures.

The short-time tensile properties of various carbon-manganese steels as determined by short-time tensile tests are given in Figs. 20 through 23. For the purpose of comparison, the properties of a plain carbon steel containing the usual percentage of manganese is included. From the proportional limit values given in Fig. 21, it will be observed that in general the values obtained with the manganese steels lie above those of the plain carbon steels.

Molybdenum and Vanadium Steels. Molybdenum and vanadium steels are both similar to chromium and so should, like chromium, increase the load-carrying ability of the alloys to which they are added. They both differ from chromium, in that smaller percentages of these elements are required to produce the same changes as in the case of the chromium alloys.



Fig. 29



Fig. 30

SUMMARY

This research was undertaken for the purpose of answering three questions: First, what relation, if any, exists between the results obtained from short-time tensile and long-time creep tests; second, what are the factors affecting the stability of metals at

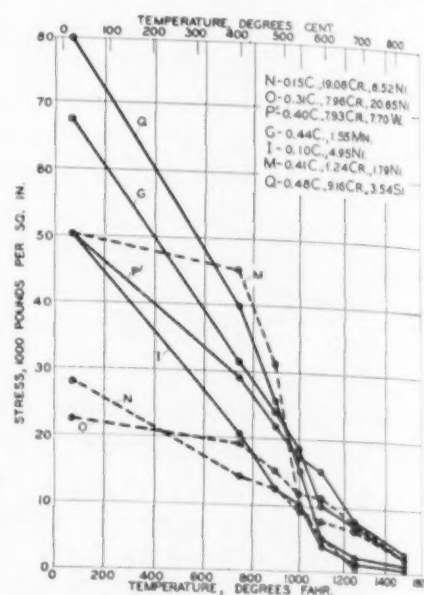
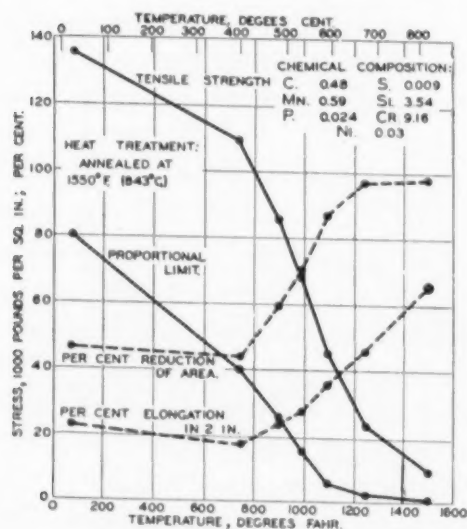


Fig. 31—Tensile Properties of Silerome at Elevated Temperatures. Fig. 32—Proportional Limit Curves for Various Selected Steels at Elevated Temperatures.

elevated temperatures; and third, what mathematical relationship exists between the variables encountered in long-time testing.

In regard to the first, it has been concluded that whether or not any relationship exists between these two forms of testing depends entirely upon the temperature range being considered. If the temperature employed is at or below the equi-cohesive temperature, which for plain carbon steels is 600 degrees Fahr. (316 degrees Cent.) or higher, then a limiting creep stress will exist and this will probably lie either at or above the proportional limit value as determined by carefully performed short-time tensile tests.

If the temperature employed is above the equi-cohesive temperature, then an absolute limiting creep stress does not exist and so there can be no relationship between the two forms of testing in this range of temperature. In this same connection, however, our findings show that stresses requiring periods of 10,000 hours or so to produce rupture in this temperature range are generally of

the same order of magnitude as the proportional limit values for the steels at these same temperatures. Therefore, carefully determined proportional limit values are useful in that they represent stresses requiring considerable periods of time to produce rupture.

In regard to proportional limit values, however, great care should be used in their selection from the literature. Due to various types of apparatus used for determining the proportional limit, values varying considerably are often reported for the same steel. For example, for a steel with which there has been occasion to do much work at this laboratory, the highest proportional limit value possible to obtain at 1000 degrees Fahr. (538 degrees Cent.) after all possible heat treatments is in the neighborhood of 15,000 pounds per square inch, while another investigator reports a value for the same steel at this same temperature of about 39,000 pounds per square inch.

Our reasons for dividing the temperature range into two parts in regard to any relationship existing between short-time and long-time tests are as follows. At temperatures below the equi-cohesive temperature, the crystalline phase is weaker than the so-called amorphous phase and the metal behaves as a perfect crystalline substance, that is, it is perfectly elastic, or nearly so, and any deformation occurs largely within the crystal. Since elastic bodies have definite proportional limits and exhibit no flow below this stress, steel, in this temperature range, will exhibit the same properties.

Above the equi-cohesive temperature, however, the so called amorphous phase becomes the weaker and the steel exhibits the properties of a viscous fluid, that is, it possesses no definite proportional limit and will flow continuously under any given stress provided sufficient time is allowed.

Considering the factors affecting the stability of metals at elevated temperatures again two separate temperature ranges must be taken into account. The stability of a metal or any other substance can only be increased by increasing the strength of or removing the weakest phase present. Since the weaker phase in steel changes at a certain temperature, it is evident that the factors affecting stability must also change at this temperature.

If the stability is desired at any temperature below the equi-cohesive temperature, the following ways are available:

- (1) The addition or formation of any element, compound, or

constituent which does not enter into solid solution with the matrix, but which by the very nature of its presence interferes with the crystal slippage, will tend to increase the stability at elevated temperatures of the material of which it forms a part.

- (2) The addition or formation of any element, compound, or constituent which tends to increase the amount of amorphous material existing between the grains of crystals in metals and like substances tends to increase the stability of the materials of which the amorphous substance forms a part.

If the stability is desired at any temperature above the equi-cohesive temperature, the following ways are available:

- (1) The addition or formation of any element, compound, or constituent which decreases the relative amount of amorphous material will increase the load-carrying ability.
- (2) The addition or formation of any element, compound, or constituent which strengthens the amorphous boundaries by interfering with plastic flow will increase the load-carrying ability.

Many alloying elements influence the stability throughout the entire temperature range of the steels to which they are added, but the effect upon the load-carrying ability may vary depending upon the temperature under consideration. Nickel for example, insofar as it is a grain refiner, increases the stability of steel below the equi-cohesive temperature but because of this same influence has the opposite effect at temperatures above this critical point. The carbide formers, as chromium, tungsten, vanadium, and manganese, increase the stability of the metals below the equi-cohesive temperatures, due to carbide particles interfering with crystalline slippage; and increase the stability above the equi-cohesive range due to the carbide particles at the grain boundaries interfering with plastic flow. Thus steels carrying carbide formers have improved stability characteristics throughout all temperature ranges, although it must be recognized that different carbide formers possess this property in varying degree.

Considering the relationship existing between the variables encountered in long-time testing, the experimental results from the expansion tests were treated mathematically to determine whether

or not stress and time could be in any way related. The equation which was found to express most accurately the results was of the following type:

$$y = a x^b$$

in which y = stress in pounds per square inch, x = duration in hours, a and b = curve constants.

ACKNOWLEDGEMENT

This paper is largely a dissertation* submitted by Claude L. Clark in partial fulfillment of the requirements for the Degree of Doctor of Philosophy in the University of Michigan. The work was done under the direction of, and in collaboration with, Professor Albert E. White.

The authors wish to express their appreciation to The Detroit Edison Company for their financial assistance given to this project. Especially do they desire to thank Mr. Hirshfeld, Chief of Research of the Detroit Edison Company, for the kindly aid and counsel which was always so graciously given.

APPENDIX

SHORT-TIME TENSILE TESTS ON VARIOUS SELECTED STEELS

Even though the results obtained by means of expansion tests show that the proportional limit is not the limiting stress below which creep will not continue until fracture occurs, they nevertheless are for the steels and temperatures considered generally of the same order of magnitude as the stresses requiring many thousands of hours to produce rupture and so are, after a fashion, a measure of the stability of the material when under sustained loads at elevated temperatures. Also, it is believed that for temperatures in the neighborhood of 1000 degrees Fahr. (538 degrees Cent.) or above, the load requiring a period of several thousand hours to produce rupture never lies much above the proportional limit value at that same temperature. For these reasons, the short-time tensile tests at elevated temperatures are valuable.

Some question arises in high-temperature testing as to just what heat treatment should be applied to the metal before subjecting it to the testing process. It has been shown several times that in all tempering operations two factors must be considered, (1) the temperature employed, and (2) the time during which the metal is maintained at that temperature. Therefore, if it is desired to subject the specimens to any sort of quenching operation, it is necessary to temper them at a sufficiently high temperature so that the time effect at the temperature at which they are to be used will not cause any marked changes in structure with corresponding changes in physical properties. Because of the necessity of following any quenching treatment with such a

*Engineering Research Bulletin, No. 11, Department of Engineering Research, University of Michigan, Ann Arbor, Michigan.

high degree of tempering, all the metals reported on in this section were subjected to either a normalizing or an annealing treatment. It is felt that these two treatments more nearly represent the working conditions that are encountered in service.

The chemical composition of the steel whose properties, as determined by the short-time tensile test, have been investigated are given in the following table.

Symbol	Class	C	Mn	Si	S	P	Cr	Ni	W
D	Carbon	0.37	0.69						
E	Manganese	0.35	1.17	0.115	0.014	0.019			
F	Manganese	0.18	1.31	0.031					
G	Manganese	0.44	1.55	0.171	0.028	0.021			
H	Manganese	0.44	2.06	0.256	0.023	0.022			
I	Nickel	0.10	0.43	0.160	0.013	0.024	0.02	4.95	
K	Nickel	0.23	0.57	0.230	0.028	0.040		4.95	
L	Chrome Nickel	0.38	0.60	0.180	0.023	0.010	0.58	1.46	
M	Chrome Nickel	0.41	0.56	0.050	0.022	0.011	1.24	1.79	
N	Chrome Nickel	0.15	0.48	0.510	0.038	0.005	19.08	8.52	
O	Nickel Chromium	0.31	0.65	0.910	0.024	0.005	7.96	20.65	
P	Chrome Tungsten	0.40	0.59	0.760	0.019	0.005	7.93		7.70
Q	Silicome	0.48	0.59	3.540	0.009	0.024	9.16	0.03	

If, however, the metals are to be used at only relatively high temperatures, such as 1000 or 1100 degrees Fahr. (538 or 593 degrees Cent.) then quenching operations will often result in greatly improved properties at these moderate elevated temperatures. The question of the effect of heat treatment upon the stability of metals at temperatures in the neighborhood of 1000 degrees Fahr. (538 degrees Cent.) will be considered in a future paper.

Short-time tensile tests were conducted on five carbon-manganese steels. The first of this group is an ordinary steel but is included so that its properties may be compared with those containing higher percentages of manganese. The results of this set are given in Figs. 20 to 23 inclusive. On the basis of proportional-limit findings, the 1.55 per cent manganese, 0.44 per cent carbon steel is the outstanding member of this group. This steel also shows the maximum reduction of area values over a large part of the temperature range. The maximum tensile strength values were obtained with the 2.06 per cent manganese, 0.44 per cent carbon steel.

The properties of two nickel steels were investigated. These both contained the same amount of nickel but the carbon content of one was approximately twice that of the other. The results obtained with these steels are shown in Fig. 24 and 25. Neither of these steels appears to possess stability at elevated temperatures as the proportional-limit values decrease very rapidly with temperature increase in both cases. Although the steel with the higher carbon content has the higher tensile strength values throughout the entire temperature range, the difference becomes very slight at the higher temperatures. Throughout the entire temperature range, these steels give proportional limit values below those obtained with the 1.55 per cent manganese steel.

The results of short-time tensile tests on two chromium-nickel steels containing relatively low percentages of chromium and nickel are given in Figs. 26 and 27. Of the two steels the one containing 1.79 per cent nickel and 1.24 per cent chromium shows superior proportional limit values at temperatures up to 1100 degrees Fahr. (593 degrees Cent.). In this range, its proportional limit values are also above those obtained with either of the nickel steels.

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Tests were also conducted on two alloys containing sufficient nickel and chromium to place the steels in the austenitic group. The results obtained with these two steels are shown in Figs. 28 and 29. Typical of all austenitic steels, these both show relatively low proportional-limit values at the lower temperatures investigated. Steels of this class show a smaller percentage of decrease in proportional limit values with increasing temperatures, however.

Silerome, a steel containing a large amount of silicon and chromium, and a relatively high chromium-tungsten steel, were also subjected to this method of testing. The results are given in Figs. 30 and 31. Silerome gave the highest proportional-limit value at atmospheric temperature of any of those tested, while the chromium-tungsten steel was superior throughout most of the remaining temperature range.

The outstanding proportional limit values in each of the above groups are shown in Fig. 32. From these curves it can be seen that 1.24 per cent chromium, 1.79 per cent nickel steel gives the highest values at 750 and 900 degrees Fahr. (399 and 482 degrees Cent.) while the 7.93 per cent chromium, 7.70 per cent tungsten steel shows the best values over the remaining higher portion of the temperature range studied.

DISCUSSION

Written Discussion: By P. G. McVetty, Research Dept., Westinghouse Electric and Mfg. Co., East Pittsburgh.

This paper indicates the large amount of time and effort that are being given to the study of the properties of metals at elevated temperatures. The designers of apparatus for high temperature service find a large and ever-increasing number of alloys available. Two main problems present themselves; first, to choose the material best suited for the purpose, second, to assign a safe working stress for this material under the conditions to be met in service. Any reliable test procedure which will simplify these two problems will be of great value.

Before accepting the conclusions reached by the authors, it is desirable to make a critical study of the basic data from which these conclusions are drawn. For example—the authors refer throughout the paper to the proportional limit as determined in the short-time tensile test. Values are given for temperatures up to 1500 degrees Fahr. even though the assumption is made that the equi-cohesive temperature for plain carbon steels is in the neighborhood of 600 degrees Fahr. I should not expect a material to exhibit elastic properties if tested above the equi-cohesive temperature. Furthermore, I doubt very much if reliable proportional limit values at elevated temperatures can be determined with the apparatus which was used for these tests.

In the long time tests, the authors admit that there is considerable uncertainty about the mathematics by which their stresses were computed. This results from the use under conditions of plastic flow of formulae based on the assumption of elasticity. Since the method of test differs from that commonly used, I should like to ask the authors, how they justify this particular type of loading for a fundamental study of this kind.

Other investigators making creep measurements, have found it desirable to go to the utmost refinement in strain measurement and temperature control. In this case strain represents a change in the circumference of the tube, and is measured as a difference between two large quantities. Measurement of a diameter is equivalent to reducing the gage length from about $3\frac{1}{2}$ inches to about 1 inch, and it implies that the tube always remains strictly circular in cross sections. On account of possible lack of homogeneity, this assumption is open to considerable question.

The use of a micrometer on a surface which has been scaled by heating introduces another error which may be of considerable magnitude. Furthermore, the tube is removed from the furnace and measured while its temperature is falling. Under these conditions it is difficult to justify the application of a refined mathematical analysis to the data obtained. In this connection, I am particularly pessimistic about the extrapolation of logarithmic curves, which are located by three or four sets of observations.

The authors have placed considerable stress upon the conditions which exist above and below the equi-cohesive temperature. This hypothesis would be strengthened considerably if photomicrographs were shown to prove that a change from fracture through the crystals to fracture between the crystals occurs in the neighborhood of 600 degrees Fahr.

Safe working loads are given for temperatures up to 1500 degrees Fahr. but the statement is made that the material will flow continuously at any stress when exposed to a temperature above 600 degrees Fahr. This appears to require some explanation of what is meant by safe working load. The values given may apply in some cases, but they surely do not apply to applications in which the deformation within the life of the material is held within close limits.

The importance of securing data on the stability of metals at elevated temperatures cannot be overestimated, but I do not agree that the methods used by the authors are suitable for the purpose.

Written Discussion: By E. R. Johnson, Central Alloy Steel Corp. We have made a number of short-time tests at elevated temperatures and find that austenitic steels will give better strength values than the ordinary steels with alpha iron. This becomes especially true when the testing temperatures exceed 1000 degrees Fahr. By high alloy content some of the austenitic steels will maintain excellent strength values at 1200 degrees Fahr. For example, the Enduro A which Prof. White used in his experiments is an alpha iron steel of 17 per cent chromium with 0.10 per cent carbon, while the corresponding austenitic steel will contain 17 per cent chromium and 8 per cent nickel. The latter develops better physical properties at a temperature of 1200 degrees Fahr. From a structural standpoint the Enduro A has two phases, alpha iron and carbides, while the austenitic steel remains a single phase throughout. Therefore there should be considerable variation of equi-cohesive temperatures.

We have also made tests of alloy steels which have been heat treated to high and low hardness limits. At the low temperatures there is a wide variation between strength values, but at a temperature of 900 to 1000

degrees Fahr. the physical properties approach each other very rapidly.

One would also think that a heat treated section with its fine grain size should have a different equi-cohesive temperature than the same steel in the annealed condition.

Authors' Written Reply to Discussion

The comments made by Mr. McVetty are much appreciated. The authors recognize that there is a difference of opinion among investigators, particularly metallurgical investigators, as to whether or not metals possess plastic or elastic properties. Also they recognize that some may hold that at certain temperatures metals possess elastic properties, while at other temperatures they become plastic in nature. We feel enough work has been done to divide metal characteristics into two groups; the one representing conditions below the equi-cohesive temperature and the other representing conditions above the equi-cohesive temperature. When above, the metals are without doubt in a plastic state, and when below, we have the two schools; one believing that the metals are in a plastic state, the other that they are in an elastic state. We recognize that the equi-cohesive temperature is not a single temperature. It varies with the composition of the metal, the state of the metal, and the conditions under which the load is applied to the metal. We mentioned 650 degrees Fahr. as being possibly the lowest equi-cohesive temperature for low-carbon steel. We recognize that metal of this composition may be in such a state that its equi-cohesive temperature may be considerably higher.

We appreciate that the methods of measuring movement in tubes as recorded in this paper are not subject to the same degree of refinement as when measurements are made in a straight tension test. The method was used because a considerable amount of material used for high-temperature purposes is tubular in character. We believe that the findings, in spite of the possible theoretical inaccuracies of the method, will be found in close agreement with the work done by other investigators under the most refined of conditions, and for that reason we feel that our mathematical deductions are worthy of consideration. We recognize that these deductions are made on a limited number of tests. We appreciate that it would be desirable to have other tests performed. We incorporated these mathematical deductions as we feel that ultimately it will be possible to convert experimental findings to mathematical laws. We therefore suggestively incorporated these laws, trusting that other investigators might be led to give consideration to this same method of attack.

We regret that Mr. McVetty does not feel that the authors used methods suitable for the investigation. Were our findings at variance with those of other investigators who are recognized as doing careful work in this field, we should feel that there might be some justification for Mr. McVetty's statement. So far as we have been able to observe, however, our findings are in very close agreement with those of other careful investigators and, therefore, we do not feel that the methods we used were unsuitable. In fact, by making these tests on tube stock, we feel that we

have added materially to the general knowledge of the subject, since our findings make it evident that long-time findings on bar material can be used with reasonable safety on tubular stock.

The authors are not able to agree with Mr. Johnson that austenitic steels will give better strength values at certain elevated temperatures than steels with alpha iron. In Fig. 32 of this paper proportional limit curves at elevated temperatures for various selected steels are given, two of these steels, namely those designated as N and O are austenitic. Up to 1000 degrees Fahr. they show values inferior to most of the other steels considered even though the other steels are in either a normalized or annealed state. At the higher temperatures they exhibit values nearly equal to the best of the alpha iron steels, but not superior to them, at least up to 1500 degrees Fahr.

We have also done considerable work upon the effect of heat treatment on the properties of various steels at temperatures in the neighborhood of 1000 degrees Fahr., the results of which are to be presented in a later paper. Here again we must disagree with Mr. Johnson as we do not find the physical properties approaching each other rapidly in this range.

THE ENGINEERING INDEX

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AIRCRAFT MANUFACTURE

METALLURGY. Metallurgy and Aeronautics. G. Py. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 30, 1928, pp. 163-164, 1 fig.

Review of papers presented before Société des Arts et Manufactures before Société des Ingenieurs Civils de France; subject matter is arranged in three chapters, first devoted to progress of metallurgy in general, second to technology of founding, mechanical working, etc., of light alloys, and third to their employment in aeronautical construction.

STEEL TUBING. Solid Drawn Tubing for Aircraft. *Flight (Lond.)*, vol. 20, no. 51, Dec. 26, 1928, pp. 1072-1074, 7 figs.

Metal construction facilitated by tube manufacturers' art; brief history of Reynolds Tube Co., Tyseley, Birmingham; details of manufacturing processes employed on solid-drawn nickel chrome-steel tube for airships similar to those used in airplane construction for highly stressed parts; tubes drawn on solid bars; hardening and tempering; higher manganese content has improved welding properties of steel; comparative tensile strengths before and after welding.

ALLOYS

AGE HARDENING. Age-Hardening in Beryllium-Copper and Low Carbon-Iron Alloys. G. Masing. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 8, 1928, pp. 189-190, 3 figs.

Summary of facts regarding age hardening of aluminum alloys and discussion on critical dispersion theory as explanation of age hardening of duralumin; investigation of beryllium-copper alloys; in regard to age

hardening and constitution, these alloys are similar to those of iron-carbon and aluminum-zinc systems; in all three cases hardening is due to decomposition of second phase. Abstract translated from Archiv fuer Eisenhuettenwesen, Sept. 1928.

EQUILIBRIUM, MODELS. Models for Analysis of Heterogenous Equilibria in Ternary Systems (Modelle zur Veranschaulichung der heterogenen Gleichgewichte in Dreistoffsystemen), R. Vogel. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 6, Dec. 1928, pp. 389-394, 12 figs.

Difficulties in interpretation of graphical representations of three-dimensional graphs are set forth; discussion of 12 new models which represent selection of simplest basic forms of ternary systems and permit a direct interpretation of three-dimensional conditions.

EUTECTIC. Anomalies of Highly Dispersed Eutectics, N. V. Aggeyer, S. A. Pogodin and N. S. Kurnakov. *Izvestiya Instituta Physico-Chimicheskovo Analiza (Leningrad)*, vol. 4, no. 1, 1928, pp. 23-38, 12 figs.

Experimental study of effect of degree of dispersion of eutectic mixtures of lead and tin and of silver and copper on their electric conductivity; hardness and coefficient of elongation. (In Russian.)

ALLOY STEEL

Progress in Alloy Steels during 1928, S. J. Hewitt. *Indus. Chemist (Lond.)*, vol. 5, no. 48, Jan. 1929, p. 18, 1 fig.

Most outstanding progress in alloy-steel trade of 1928 is adapting of tungsten carbide for high-speed cutting tools and introduction of nitrogenization process for hardening steels; important development in technique of high-frequency melting of steels and steel alloys.

Trends in Engineering Alloy Steels, C. E.

Those members who are making a practice of clipping items for filing in their own filing system may obtain extra copies of the Engineering Index pages gratis by addressing their request to the society headquarters, whereby their names will be placed on a mailing list to receive extra copies regularly.

MacQuigg. *Iron Age*, vol. 123, no. 5, Jan. 31, 1929, pp. 333-334.

Some present trends in alloy steels; lower carbon and higher alloy content insure adequate strength with greater toughness and ductility; simpler heat treatments possible; silicon and manganese steels have certain advantages; balance between carbon and alloy content; three per cent chromium steel low in carbon. Abstract of paper presented before Am. Soc. Steel Treating.

STRENGTH—The Relative Safeties of Mild and High-Tensile Alloyed Steels under Alternating and Pulsating Stresses, B. P. Haigh. *Chem. and Industry (Lond.)*, vol. 48, no. 2, Jan. 11, 1929, pp. 23-30, 8 figs.

Reviews and discusses wide variety of tests and experiments with alternating and pulsating stresses to illustrate and contrast conditions in which fatigue cracking may be source of danger; endurance and fatigue limit; methods of measuring fatigue strength; comparisons between fatigue and tensile strengths; graphic comparison of different steel; factors of safety.

ALLOY STEEL CASTINGS

HEAT TREATMENT. Heat Treatment of Carbon and Alloy Steel Castings, J. E. Donnellan. *Iron Age*, vol. 123, no. 5, Jan. 31, 1929, pp. 342 and 344.

Problems in quenching alloy steel castings; double normalizing treatment of carbon steel castings; alloy castings should, in most cases, be tempered after normalizing, to relieve brittleness and strain caused from air quench and also to increase ductility and impact values; avoiding warping and straightening; metallography of two steels. Abstract of paper presented before Western Metal Congress.

ALLOY STEEL PRODUCTS

Production of Alloy Steel Ingots Increases 97 Per Cent. *Iron Trade Rev.*, vol. 84, no. 1, Jan. 3, 1929, pp. 44-45.

Progress in alloy-steel industry 18 times as rapid as in general steel industry during past 15 years; automotive consumption increases; of 43,311 tons of cold-drawn alloy bars produced in 1928, automotive industry absorbed 30,970 tons; of 49,026 tons of pipe and tubes made last year, automotive and airplane industries purchased 44,514 tons; output of forgings of alloy quality amounted to 14,602 tons.

ALUMINUM

CASTING. Some Difficulties Encountered in Aluminum Founding. *Iron and Steel of Canada (Gardenvale, Que.)*, vol. 11, no. 12, Dec. 1928, pp. 387 and 389.

Obstacles that have to overcome resemble, in their character, those encountered in molten steel, chill and mixed casting methods; different types of aluminum castings require special treatment.

ALUMINUM ALLOYS

PROTECTIVE COATINGS. Heat-Insulating Properties of Aluminum and Their Applications (Les propriétés calorifiques de l'aluminium et leurs applications), M. de Biran. *Chaleur et Industrie (Paris)*, vol. 9, no. 103, Nov. 1928, pp. 413-417.

Résumé of certain principles relating to heat transmission; aluminum foil and coatings; aluminum painting of furnaces, chim-

neys, gas pipe and producers; painting of heat-insulated surfaces; heat insulation of small tubes.

STRENGTH. Strength of Crystals of Technical Aluminum Alloys (Festigkeitseigenschaften von Kristallen einer veredelbaren Aluminiumlegierung), G. Sachs. *Zeit. fuer Metallkunde (Berlin)*, vol. 20, no. 12, Dec. 1928, pp. 428-430, 7 figs.

Investigation of individual crystals of technically important alloys; tensile tests, X-ray analyses, and recrystallization tests.

ALUMINUM-SILICON ALLOYS

Aluminum-Silicon Alloys: Their Properties and Some Applications, R. B. Deeley. *Foundry Trade J. (Lond.)*, vol. 40, nos. 647 and 648, Jan. 10 and 17, 1929, pp. 29-32 and 49-51, 14 figs. See also *Metal Industry (Lond.)*, vol. 34, nos. 3 and 4, Jan. 18 and 25, 1929, pp. 75-78 and 103-106, 6 figs.

Modification treatment, which improves properties of aluminum-silicon alloys, consists of treating alloys containing from 5 to 20 per cent of silicon, in molten state, before casting with alkaline fluoride, of mixture containing alkaline fluoride with alkaline chloride, or even substance which under this treatment would yield alkaline fluoride. Paper presented before Inst. Brit. Foundrymen and Inst. of Metals.

ANTIMONY ALLOYS

Some Antimony Alloys. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 28, 1928, pp. 182-184, 4 figs.

Review of three papers recently appearing in *Revue de Métallurgie*; two of them are by W. Broniewski and L. Sliwowski, and deal, first with antimony lead and second with antimony-tin systems, while remaining one is translation of paper by Dean, Zickrick, and Nix; all three have in common that they rely to very considerable extent on physical property measurements, principally electrical conductivity, for elucidation of diagram.

AUTOMOBILE MANUFACTURE

HEAT TREATMENT. Heat Treatment of Automotive Parts, J. M. Watson. *Iron Age*, vol. 123, no. 5, Jan. 31, 1929, p. 344.

Alloy steels for lighter parts; properties desired in automotive parts vary with duties they are to perform and heat-treating processes selected are arrived at in accordance with use of part; modern heat-treating furnaces of continuous type; mass heat treatment with laboratory precision. Abstract of paper presented before Western Metal Congress.

HEAT TREATMENT. Heat-Treating at the Reo Plant. *Am. Mach.*, vol. 79, no. 5, Jan. 31, 1929, pp. 196-197, 8 figs.

Heat-treatment methods employed at plant of Reo Motor Car Co. are illustrated in eight halftones, each accompanied by brief description.

AUTOMOBILES

GEARS AND GEARING — HARDENING. Continuous Furnace Improves Gear Hardening Process. *Automotive Industries*, vol. 59, no. 25, Dec. 22, 1928, pp. 904-908, 4 figs.

Practice of Warner Gear Co. is discussed; upsetting for all gears insisted upon; forging temperature; ideal way of annealing forgings en masse is to have them pass through

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series of heating and cooling zones maintained uniform by automatic pushing and fuel control; Hump furnace method best but lot of electrical, metallurgical, and mechanical attention necessary; cyaniding methods; eliminating high quench; best quenching temperatures. Abstract of paper presented before Am. Soc. Steel Treating.

BERYLLIUM

ELECTROLYTIC PRODUCTION. The Production and Uses of Beryllium, K. Illig. *Metal Industry (Lond.)*, vol. 33, no. 26, Dec. 28, 1928, pp. 614-617.

Although Wohler obtained metallic beryllium 100 years ago by reduction with potassium it was not until recently that Stock and Goldschmidt produced metal by fused-electrolyte method; their method has been developed by Beryllium Institute, and large masses of metal (150 g.) are now turned out commercially; properties of certain new metal alloys are enumerated. Paper read before Am. Electrochem. Soc.

BERYLLIUM ALLOYS

Change in Length and Modulus of Elasticity of Beryllium-Copper Alloys during Heat Treatment (Änderungen der Laenge und des Elastizitätsmodulus der Beryllium-Kupferlegierungen bei der Verguetung), O. Dahl and C. Haase. *Zeit. fuer Metallkunde (Berlin)*, vol. 20, no. 12, Dec. 1928, pp. 433-436, 5 figs.

Change in both of these properties is not due to refining process, but is directly influenced by degree of separation.

X-Ray Analysis of Improvement of Beryllium Alloys (Die Verguetung bei den Beryllium-Kupferlegierungen im Roentgenbild), O. Dahl, E. Holm and G. Masing. *Zeit. fuer Metallkunde (Berlin)*, vol. 20, no. 12, Dec. 1928, pp. 431-433, 3 figs.

Results of investigation carried out at research laboratory of Siemens-Werke; in X-ray examination of heat treatment wires were quenched from 800 deg. Cent. and aged at different temperatures.

BOILER PLATES

AGE HARDENING. Age Hardening of Boiler Plate and its Prevention (Alterungserscheinungen an Dampfkesseln und ihre Vermeidung), F. Nehl. *Zeit. des Bayerischen Revisions-Vereins (Munich)*, vol. 32, nos. 23 and 24, Dec. 15 and 31, 1928, pp. 315-317, 6 figs. and 324-325, 1 fig.

Causes and effect of aging are discussed and two means of prevention are recommended; first is use of steel less subject to aging; cost of such material, however, is often prohibitive; by use of welded drums of standard steel, it is possible to obtain boiler material entirely resistant to aging.

BRASS

ALUMINUM EFFECT. The Effect of Aluminum on Brasses, J. Silberstein. *Metal Industry (Lond.)*, vol. 34, no. 1, Jan. 4, 1929, pp. 5-6.

Explanation of behavior of aluminum in brasses and bronzes, and statement of cases in which addition of aluminum is of advantage and in which it is not; melting loss influenced by aluminum; suitability of aluminum bronzes for permanent-mold castings; copper base alloys strengthened by addition

of aluminum; harmful effects of aluminum addition.

CADMIUM PLATING

The Electro-Deposition of Cadmium and Its Alloys. *Am. Metal Market*, vol. 35, no. 240, 2nd section, Dec. 15, 1928, pp. 18-21 and 51.

Deposit of cadmium had color more nearly approaching that of silver; cadmium plate adheres tenaciously and takes equally well on all metals; cadmium deposition from cyanide solutions, from fluosilicate, perchlorate, fluoborate and fluoride solutions, from sulphate, phosphate, chloride, bromide, iodide and acetate solutions, and from ammoniacal and alkaline tartrate solutions; electrodeposition of cadmium-silver alloys.

CASE HARDENING

Case Hardening of Steels by Special Manganese alloys (Sur quelques cementations d'aciers par des alliages speciaux à base de manganese), J. Cournot. *Revue de Métallurgie (Paris)*, vol. 25, no. 12, Dec. 1928, pp. 669-670.

Composition of powder, structure obtained and hardness realized; resistance to corrosion.

BATHS. Salt Baths and Tool-Handling, C. B. Gordon-Sale. *Can. Machy. (Toronto)*, vol. 39, no. 25, Dec. 13, 1928, pp. 34 and 66-67.

Salt baths not always adaptable for certain work and seldom used for hardening high-speed steels; type of pot and heating method described; mixtures for temperatures ranging from 240 to 500 deg. Cent.; cyanide baths; advantages of salt bath over lead one.

CYANIDE BATH. Salt Bath Hardening. *Automobile Engr. (Lond.)*, vol. 19, no. 250, Jan. 1929, p. 16.

New flux for rapid and deep carburizing, known as Durferrit hardening flux, is described; by addition of various ingredients to sodium cyanide Deutsche Gold- und Silber-Scheideanstalt, of Frankfurt-on-Main, have flux evolved by means of which satisfactory case up to 1 mm. depth can be obtained; good finish of work after treatment; absence of distortion claimed.

CYANIDE BATH. The Cyanide Bath, V. E. Hillman. *Fuels and Furnaces*, vol. 7, no. 1, Jan. 1929, pp. 31-34.

Low-carbon steel, treated by sprinkling potassium cyanide or its equivalent over heated metal or immersing parts in molten solution rich in cyanogen content, develops very hard case in newly formed layer; ventilation necessary during operation of bath, its composition, operation, and resultant case and core; various compounds that are used in making up bath, their concentration and proper method of keeping them up to strength, proper temperature at which to operate cyanide bath, depth of case obtained, and physical qualities of both case and core.

DISTORTION. Distortion in Case-Hardening, M. R. Barat. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 28, 1928, pp. 181-182.

Trouble due to distortion during case-hardening and quenching is capable of more satisfactory treatment than by subsequent rectification; author suggests that it may be possible to predict and calculate amount and nature of distortion likely to occur and to

provide for it, to very great extent, by changing size and shape of object as produced prior to cementation; causes of distortion. Abstract translated from *Revue de Métallurgie*, Oct. 1928.

CASTING

CENTRIFUGAL. A Notable Foundry. *Iron and Steel Industry (Lond.)*, vol. 2, no. 3, Dec. 1928, pp. 69-71, 3 figs.

Centrifugal casting department of Thorncliffe Iron Works of Newton and Chambers and Co., is described; two large cupolas and one small experimental cupola; production of piston-ring drums and cylinder liners; discovery of means whereby low-silicon low-total-carbon-content material can be cast by centrifugal casting process; Hurst-Ball machine.

CAST IRON

ALLOY. Alloy Cast Irons, A. B. Everest. *Foundry Trade J. (Lond.)*, vol. 40, no. 648, Jan. 17, 1929, pp. 45-48, 2 figs.

Ultimate structure of cast iron is determined primarily by relative stability of carbide and graphite in iron, and thermal history of sample of iron under consideration; influence of composition; choice of alloy element; influence of aluminum; influence of nickel in cast iron; use of nickel-chromium; influence of chromium; cost of alloy additions; nickel in special cast iron.

GRAPHITE STRUCTURE. Three-Dimensional Structure of Graphite (Die Raumform des Graphits), F. Roll. *Gießerei (Duesseldorf)*, vol. 15, no. 51, Dec. 21, 1928, pp. 1270-1274, 16 figs.

Method is described by which determination of volumetric form of graphite by mechanical means can be effected; catalytic investigations are also briefly described. Bibliography.

HEAT TREATMENT. Heat Treatment of Cast Iron, J. W. Donaldson. *Foundry Trade J.*, vol. 39, no. 644, Dec. 20, 1928, pp. 445-446.

Discussion of paper which appeared in Oct. 25 and Nov. 1 issues of journal. Critical silicon percentage; results of phosphorus experiments showing stability effects reconsidered, in case that specimens were heated above critical temperature; author questioned as to whether preliminary annealing at higher temperature prevents growth when material is again heated to lower temperature.

NICKEL. The Influence of Nickel on Combined Carbon in Gray Iron, J. R. Houston. *Am. Soc. Steel Test.—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 145-153, and (discussion) 153-157 and 169, 10 figs.

In recent years claim has been made for nickel that it is pearlite builder in gray iron, and that by its use in favorable composition, both amount and structural form of combined carbon may be controlled to marked degree, thereby giving density to heavy sections and machinability to light sections, which, if true, makes it valuable tool in hands of foundry metallurgist.

NICKEL AND CHROMIUM CONTENT. The Influence of Nickel-Chromium on Cast Iron, A. B. Everest and D. Hanson. *Foundry Trade J. (Lond.)*, vol. 40, no. 646, Jan. 3, 1929, pp. 5-10, 14 figs.

Influence of nickel and chromium in combination and individually on foundry iron;

laboratory tests on influence of nickel and chromium on high-grade cylinder iron; properties of series of step castings; chill; hardness determination; fractures; microstructure; nickel will harden gray portions of cast iron and eliminate its tendency to chilling; chromium increases hardness of iron and also its tendency to chill; practical workshop tests being made.

PROPERTIES. Investigation of Relation Between Compressive and Bending Strength of Cast Iron (Beitrag zur Untersuchung des Zusammenhanges zwischen Druck und Biegefestigkeit des Gusseisens), St. Nadassan. *Gießerei (Duesseldorf)*, vol. 15, no. 50, Dec. 14, 1928, pp. 1251-1253, 5 figs.

Results of tests show relationship between pressure and bending strength of manganese-rich cast iron; tests also show that manganese content of over 6 per cent does not influence strength.

PROPERTIES. Improves Cast Gray Iron With Steel Scrap Additions, E. E. Marbaker. *Foundry*, vol. 57, no. 1, Jan. 1, 1929, pp. 10-12, 3 figs.

Methods of securing desirable structures in cast iron further discussed; features of Coralli cupola and Dechesne jolting process; jolting molten metal in ladle also found beneficial since gases are freed, sulphur is made innocuous and solution is mixed thoroughly; tables give results of tests of cast iron made by various methods. (Concluded.)

CHAINS

ANNEALING. Service Annealing of Sling and Crane Chains, W. J. Merten. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 2, Feb. 1929, pp. 193-209 and (discussion) 209-216, 21 figs.

Inconsistent results in ductility and strength are ordinarily obtained when chains are annealed at temperatures unsuited for links that have been severely deformed and cold worked in service; author recommends uniform recrystallization at temperatures considerably above transformation range; calculation of permissible stresses based on average cross-sectional dimensions only.

WROUGHT IRON—FAILURE. The Causes of Failure of Wrought-Iron Chain and Cable, H. J. Gough and A. J. Murphy. *Instn. Mech. Engrs.—Proc. (Lond.)*, no. 2, 1928, pp. 293-326 and (discussion) 327-352, 29 figs.

Failures and characteristics of overheated wrought iron; failures due to faulty welds or to welds which have deteriorated in service; series of experiments made to study effect of repeated static straining; investigation of properties of various wrought-iron chains drawn from service of known history; conclusions drawn from results of research.

CHROMIUM ALLOYS

Chromium Alloys, F. M. Becket. *Min. and Met.*, vol. 10, no. 265, Jan. 1929, pp. 10-15, 1 fig.

Types of chromium steel; chromium supplements and enhances effect of carbon in hardening; its effect on corrosion resistance. (Concluded.)

CHROMIUM-COPPER STEEL

Chromium-Copper Steels as Possible Corrosion Resisting Ferrous Alloys. *Am. Soc.*

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for Steel Treating.—*Trans.*, vol. 15, no. 1, Jan. 1929, pp. 36-48, 17 figs.

Data presented are results of experimental laboratory work in study of merits of chromium-copper combination in steel; study was restricted to commercial and practical ranges of the two elements; main purpose of study was development of commercial steels of high physical properties combined with fair corrosion resistance and producible at commercially reasonable cost; results are sufficient indication of trend of qualifications of material.

CHROMIUM-NICKEL STEEL

HARDNESS. Hardness and Impact Resistance of Chromium-Nickel Steel, B. F. Shepherd. *Can. Machy. (Toronto)*, vol. 39, no. 26, Dec. 27, 1928, pp. 213-214.

Results of Izod impact, hardness, and tensile tests of chromium-nickel steel of S.A.E. 3250 type with varying nickel and carbon content are discussed; tests indicate that this steel should be used in standard analysis range; tempering temperature of 550 deg. Fahr. should be used to obtain as great resistance as possible to eccentric loading conditions where hardness is not extremely important. Paper presented before Am. Soc. Steel Treating.

CHROMIUM PLATING

A Study of Chromium Plating, R. Schneidewind. *Univ. of Mich.—Eng. Research Bul.*, no. 10, Sept. 1928, 140 pp., 85 figs.

Non-technical account of electro-deposited chromium and commercial methods for its preparation; detailed account of original research done at University of Michigan during years 1927 and 1928; review of scientific and patent literatures on subject; summary of review of literature arranged in tabular form.

Chromium Plating. *Indus. Power*, vol. 11, no. 2, Feb. 1929, pp. 159-162, 164, 166 and 168, 5 figs.

Review of chromium characteristics; electroplating baths; automatic temperature-control devices; how bright deposits are obtained.

CHROMIUM STEEL

METALLURGY. Study of Ternary Chrome Steels (Contribution à l'étude des aciers ternaires au chrome), J. Pomey and P. Voulet. *Revue de Metallurgie (Paris)*, vol. 25, no. 12, Dec. 1928, pp. 665-667, 3 figs.

Martensitic and austenitic steels are studied and location of A points discussed.

COPPER ALLOYS

PROPERTIES. Content and Uses of Brass and Other Copper Alloys, W. H. Bassett. *Can. Machy. (Toronto)*, vol. 40, no. 1, Jan. 10, 1929, pp. 26-32, 8 figs.

Description of various copper alloys; Lake copper; nature of refined and deoxidized copper; deoxidizing effect of boron suboxide; uses for cold-rolled and annealed copper; effect of silver; Everdur; copper in engineering; market brass; hot working; leaded brass; German silver; use of copper and copper alloy rapidly increasing.

COPPER-BERYLLIUM ALLOYS

Copper-Beryllium Alloys. *Metallurgist (Supp. to Engineer, Lond.)*, Jan. 25, 1929, pp. 13-15, 2 figs.

Review of two papers published in Dec. issue of *Zeit. fuer Metallkunde*; first paper, by Dahl, Holm and Masing, deals with X-ray study of age-hardening process in these alloys; second paper, by Dahl and Haase, deals with changes of length and of elastic modulus in copper-beryllium alloys during age hardening.

COUPLINGS, FLEXIBLE

FATIGUE. What Happens in Couplings; Fatigue of Materials; What It Is and How It Occurs in Flexible Couplings, H. F. Moore. *Iron and Steel Engr.*, vol. 5, no. 12, Dec. 1928, pp. 513-517, 8 figs.

Fatigue of metals discussed; dangerous effect of reversing stresses due to misalignment of shafting; function of flexible coupling; table gives fatigue limit and energy storing capacity for various metals; coupling with flexible materials eliminated; Fast's coupling developed on floating-sleeve principle with oil as load-carrying agent.

DIE CASTING

Die Casting Machines Make Variety of Parts, H. R. Simonds. *Foundry*, vol. 57, no. 2, Jan. 15, 1929, pp. 72-77, 9 figs.

Details of plant and methods of Newton Die Casting Corp., New Haven, Conn.; flow of material through plant; die-casting machines and operations; method of cooling dies; scheme of discarding old dies; finishing operations; cost estimating.

ALLOYS. Zinc Base Die Casting Alloys, R. M. Curtis. *Sibley JI. of Eng.*, vol. 42, no. 9, Dec. 1928, pp. 318-320 and 346, 9 figs.

Discussion of important facts and developments pertaining to zinc-base die-casting alloys; briefly outlines both importance and process of die casting; summarizes best zinc base die-casting alloys available today.

DIES, FORGING

MANUFACTURE. The Manufacture of Die Blocks, J. A. Succop. *Heat Treating and Forging*, vol. 15, no. 1, Jan. 1929, pp. 51-54, 4 figs.

Making of die blocks is traced from open-hearth charge to heat-treated and tested product; steel is made in two 18-ton acid open-hearth furnaces; acid process is used; forging and heat-treating practice; factors in proper heat treatment of die blocks; proper heat-treating temperatures, uniformity of temperature and time on temperature; since hardening cracks are occasionally formed, die is subjected to ringing test; chemical composition and physical properties.

DROP FORGINGS

Producing an Unusual Drop Forging, W. Hopson. *Heat Treating and Forging*, vol. 15, no. 1, Jan. 1929, pp. 60-61, 3 figs.

Method whereby piece could be produced as drop forging, requirements of which were that teeth had to be sharp and of even height, so that when toothed surface was placed on piece of 6-in. pipe, every tooth came in contact with pipe.

ELECTRIC FURNACES

ANNEALING. Electric Annealing Furnaces (Der elektrische Temperofen), F. Russ. *Gieserei (Duesseldorf)*, vol. 16, no. 3, Jan. 18, 1929, pp. 61-62, 3 figs.

Advantages of electric annealing are set

forth, and two types of furnaces are recommended; special electric furnace is described which is equipped with hood and has annular muffle with heating elements arranged along walls; results obtained with this furnace are given.

ANNEALING. Electric Furnace Anneals 5.4 lb. of Steel per Kw.-hr. *Elec. West*, vol. 62, no. 1, Jan. 1, 1929, p. 10, 3 figs.

Description of furnace 5 ft. 6 in. wide, 11 ft. deep and 5 ft. 6 in. high, inside measurements; temperature of furnace is controlled automatically; furnace has demand of 274 kw.; outstanding advantages noted by American Foundry & Machine Co. in electric-furnace operation.

HEAT TREATING. Electric Heat Treating Proves 70 Per Cent Lower in Cost, H. E. Scarbrough. *Elec. World*, vol. 93, no. 3, Jan. 19, 1929, p. 154.

Vertical, cylindrical-type furnace, 30 in. in diam. by 30 in. deep, was installed about year ago at Maytag Co., Newton, Iowa, and has proved both economical and satisfactory; it is rated 37 kw. 220 volts, single phase; used for normalizing and annealing small crankshafts used in manufacture of gasoline engine-driven washing machines.

HEAT TREATING. Use of Electricity in the Heat Treatment of Metals, C. L. Ipsen. *Iron Age*, vol. 123, no. 5, Jan. 31, 1929, p. 344.

Advantages of electric furnaces for heat treating are discussed. Abstract of paper presented before Western Metal Congress.

HIGH FREQUENCY. Spark Gaps in High-Frequency Furnaces (Les éclateurs de fours à haute fréquence), R. Dufour. *Revue Générale de l'Electricité (Paris)*, vol. 24, no. 26, Dec. 29, 1928, pp. 970-972, 3 figs.

General discussion of spark gaps, kinds and power limitations.

INDUCTION. Electric Induction Furnaces "Russ" Type (Fours électrique à induction système "Russ"). *Jl. du Four Electrique (Paris)*, vol. 37, no. 12, Dec. 1928, pp. 369-374, 9 figs.

Comparison of large and small induction furnaces; advantages; installations and operation.

INDUCTION. Heating by Induction (Le chauffage par induction), P. Bunet. *Revue Générale de l'Electricité (Paris)*, vol. 24, no. 26, Dec. 29, 1928, pp. 965-970, 4 figs.

General discussion on Kelvin method; furnaces fed by alternators, by discharge from condensers; arrangement of windings; use of iron in furnaces. Bibliography.

METALLURGICAL. The Electric Furnace in the Iron Industry, J. G. Webb. *Stone and Webster Jl.*, vol. 44, no. 1, Jan. 1929, pp. 56-62, 3 figs.

There are three types of electric furnaces, induction, resistance, and arc furnace; application of electric furnace in manufacture of specialties is illustrated by example.

STEEL MELTING. Electric Melting of Alloy Steels, H. M. German. *Iron and Steel Engr.*, vol. 5, no. 12, Dec. 1928, pp. 507-510.

Stassano, Kjellin, Heroult, and Girod electric furnaces described in historical sketch; growth of electrical steel industry in United States; cold melting practices with complete, partial, and no oxidation; selection of method governed by analysis of available charging materials; refining slags; temperature of metal before tapping. Paper presented be-

fore Industrial Heating Conference at Penn State College.

TEMPERING. Rejects Few in Electric Spring Tempering, P. N. Rugg. *Elec. World*, vol. 93, no. 5, Feb. 2, 1929, pp. 251-252, 2 figs.

Electric conveyor-type tempering oven rated at 10 kw. built by Barber-Colman Co., Framingham, Mass., for use in production of small steel springs of about 0.40 per cent carbon component formerly tempered by gas.

FERROALLOYS

CASE HARDENING. Depth and Character of Case Induced by Mixtures of Ferro-Alloys with Carburizing Compounds, E. G. Mahin and R. C. Spencer. *Am. Soc. Steel Test—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 117-141 and (discussion) 141-144, 14 figs.

Experiments described show that if silicon absorption is confined to surface layers of iron or steel, carbon adsorption during case carburization may be retarded without materially affecting rate of inward migration of carbon after it has been absorbed; by utilizing this principle it is possible to produce deep case without zone of free cementite by using ordinary carburizer in conjunction with ferrosilicon.

FORGINGS

MACHINE. Some Notes on the Quality of Upset Machine Forgings, E. R. Frost. *Can. Machy. (Toronto)*, vol. 39, no. 26, Dec. 27, 1928, pp. 202-205, 20 figs.

Upsetting of steel in forging machines is discussed; results of tests and photomicrographs are given to show endwise working of material places grains in closer endwise formation; other tests described; table of average elastic limit, ultimate strength elongation, and reduction of area; no evidence exists to indicate any unsatisfactory results through upsetting on any alloy steels or nonferrous metals. Paper presented before Am. Soc. for Steel Treating.

FURNACES, ANNEALING

CONTINUOUS. Double Deck Conveyor Furnaces Used for Annealing Stampings. *Fuels and Furnaces*, vol. 6, no. 12, Dec. 1928, pp. 1655-1657, 3 figs.

Oil-fired furnaces, of continuous conveyor hearth type in which product is heated in upper chamber and cooled in lower chambers are used for annealing of automobile front-wheel hub stampings between drawing operations.

GAS FIRED. Gas Fired Car Bottom Furnaces Used for Annealing Steel Castings, W. P. Hand. *Fuels and Furnaces*, vol. 6, no. 12, Dec. 1928, pp. 1751-1752, 3 figs.

Carbon and manganese steel parts accurately heat treated; furnaces automatically controlled; construction and operation of furnace; construction of car.

FURNACES, FORGING

FUELS. Test of Heating of Continuous Furnace by Mixture of Blast-Furnace Gas and Pulverized Coal with Possibility of Continuing the Heating with Pulverized-Coal Alone (Essais de chauffage d'un four continu par mélange de gaz de hauts-fourneaux et de charbon pulvérisé, avec possibilité de continuer le chauffage au charbon pulvérisé seul).

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G. Rey. *Chaleur et Industrie* (Paris), vol. 9, no. 162, Oct. 1928, pp. 117-131 and (discussion) 131-132, 12 figs.
Description of tests on continuous furnace in forge of Creusot Works; air necessary for combustion; test made and results obtained.

FURNACES, HEATING

CONTINUOUS. Triple-Fired Heating Furnace Has Many Features of Design. *Iron Trade Rev.*, vol. 84, no. 3, Jan. 17, 1929, pp. 200-202, 4 figs.

Details of two large continuous recuperative furnaces especially designed for reheating high-quality alloy-steel blooms on mass-production basis, in new mill of Timken Steel and Tube Co., Canton, Ohio; three zones incorporated in hearth construction afford heating semi-finished material at uniform temperatures and provide flexibility in tonnage output; natural gas used for fuel; capacity 40 tons an hour.

CONTINUOUS. Reheating Furnaces for Alloy Blooms. *Iron Age*, vol. 122, no. 26, Dec. 27, 1928, pp. 1632-1634, 3 figs.

Reheating high-quality alloy blooms on quantity production basis at new mill of Timken Steel and Tube Co., Canton, Ohio, is described; two large continuous reheating triple-fired recuperative furnaces using natural gas and each having capacity of 40 tons per hour; soaking chamber forms part of furnaces; hearth has three zones; control to be made automatic; refractory recuperator under furnace.

CONTINUOUS. Continuous Furnaces for Reheating Alloy Blooms. *Fuels and Furnaces*, vol. 7, no. 1, Jan. 1929, pp. 55-58, 4 figs.

Alloy blooms are reheated in two triple-fired recuperative-type reheating furnaces, each furnace having capacity of 40 tons per hour; hearth construction; system of firing; recuperation; operation of furnace.

REGENERATORS. Study of Reheating Furnaces with Recuperators in Which Part of the Heating Fumes are Reversed (Etude des fours à réchauffer, avec recuperateurs, dans lesquels on détourne une partie des fumées chauffantes), J. Seigle. *Genie Civil* (Paris), vol. 94, no. 1, Jan. 5, 1929, pp. 7-12, 11 figs.

Study of heating and recuperating process; thermal efficiency of single furnace using progressive heating with no loss by radiation; operation of recuperative furnaces.

FURNACES, INDUSTRIAL

DESIGN. Practical Industrial-Furnace Design. M. H. Mawhinney. *Mech. Eng.*, vol. 51, no. 1, Jan. 1929, pp. 93-94.

Review of book published by J. Wiley and Sons, Inc.; brief historical review of furnace development; selection of fuels, their characteristics and comparative costs; methods of heat application and furnace types embodying them; refractory design and construction; design of metal parts of furnace proper; temperature measurement and control.

FURNACES, METALLURGICAL

PULVERIZED COAL. New types of Pulverized-Coal-Fired Furnaces in Steel Plants (Neuere Bauarten kohlenstaubgefeuerter Oefen auf Huetttenwerken), G. Kehren. *Stahl und Eisen* (Duesseldorf), vol. 48, no. 51, Dec. 20, 1928, pp. 1769-1773 and (discussion) 1773-1775, 8 figs.

Conditions governing quality required of

pulverized coal are discussed, and design of different reheating furnaces and welding furnaces with pulverized-coal firing are described.

HEAT TREATING EQUIPMENT

Shop Equipment Review—Heat-Treating Equipment. *Am. Mach.*, vol. 70, no. 3, Jan. 17, 1929, pp. 101-103, 10 figs.

Semi-annual résumé of heat-treating equipment described in Shop Equipment News Section of magazine during last six months of 1928; details of electric furnaces, heat exchanger, annealing pot, and furnace pot.

HEAT TREATMENT

TEMPERATURE CONTROL. Scientific Control of Heating in Heat Treatments (Le controle scientifique du chauffage dans les traitements thermiques), R. Gaillard. *Métallurgie* (Paris), vol. 60, no. 48, Nov. 29, 1928, pp. 27-30, 5 figs.

Discussion on scientific research in heat treating and necessity for scientific control of temperature and its importance.

HIGH SPEED STEEL

HARDENING. Hardening High-speed Steel, C. B. Gordon-Sale. *Machy.* (London), vol. 33, no. 843, Dec. 6, 1928, pp. 306-307, 2 figs.

New practical method developed for purpose of utilizing advantages of steel for special tools or irregular section where it was necessary to finish tool before hardening; tools packed in nichrome or aluminized steel box with good-quality charcoal, box heated to 1050 deg. Cent. for 3 or 4 hr., and tools quenched in oil.

INGOT MOLDS

Heat Balance Between Ingot and Mold Wall (Ueber den Waermeausgleich zwischen Block und Kokillenwandung), B. Matuschka. *Archiv fuer das Eisenhuetttenwesen* (Duesseldorf), vol. 2, no. 7, Jan. 1929, pp. 405-411 and (discussion) 411-413, 13 figs.

Heat movement occurring with solidification of ingots in molds is investigated, and it is shown that that part of solidification which results from heat balance between ingot and mold wall is deciding factor in formation of crystalline structure; temperature curves of ingot and mold walls are plotted and important thermotechnical influence of ingot removal is shown.

IRON AND STEEL

CORROSION. Colloidal Theory of Rusting Process (Ueber die Kolloidtheorie des Rostvorganges), G. Schikorr. *Korrosion und Metallschutz* (Berlin), vol. 4, no. 11, Nov. 1928, pp. 242-245, 1 fig.

Report of Dahlem Government Bureau for testing materials; criticism of Newton Friend's theory; author supports old theory.

CORROSION PREVENTION. Effect of Additions of Lime and Soda Ash to Brackish Water on the Corrosion of Iron and Steel, H. O. Forrest, J. K. Roberts and B. E. Roetheli. *Indus. and Engr. Chem.*, vol. 21, no. 1, Jan. 1929, pp. 33-35, 6 figs.

Treatment of brackish or semi-salt water to retard corrosion of exposed iron or steel structures is studied; larger-scale studies are recommended to determine minimum quantity of lime or soda ash necessary to maintain thin protective scale on structure.

OXYGEN DETERMINATION. Present-Day Methods of Determining Oxygen Content of Steel and Iron (Ueber den heutigen Stand der Bestimmung von Sauerstoff in Stahl und Eisen), O. Meyer. *Zeit. fuer angewandte Chemie (Berlin)*, vol. 41, nos. 48 and 49, Dec. 1 and 8, 1928, pp. 1273-1276 and 1295-1298, 6 figs.

Discussion of residue and reduction methods; review of recent researches in Germany, Sweden, Great Britain, and United States.

SLAG INCLUSIONS. Present Knowledge of Slag Inclusions in Iron and Steel (Unsere gegenwaertigen Kenntnisse der Schlackeneinschluesse in Eisen und Stahl), C. Benedicks and H. Lofquist. *Stahl u. Eisen (Duesseldorf)*, vol. 48, no. 52, Dec. 27, 1928, pp. 1828-1829.

Review, by W. Hessenbruch, of first part of work presented at International Congress for Materials Testing at Amsterdam, Sept. 1927; analysis of inclusions and their effects on properties of steel; behavior of inclusions in charge before solidification; influence of slag particles on grain size of metal. Original work contains extensive bibliography, which, in opinion of reviewer, is one of most important features of work.

IRON AND STEEL RESEARCH

Steel Owes Much to Research, G. K. Burgess. *Iron Age*, vol. 123, no. 1, Jan. 3, 1929, pp. 40-41.

Notable progress in alloy steels in past decade; advantages linked with research; several new tools research developed; malleable and cast iron made over; various advances briefly catalogued; group method of study developed.

MANGANESE STEEL

CASTING. Explains Manganese Steel Production in the Electric Furnace, J. Trantin, Jr. *Foundry*, vol. 57, no. 2, Jan. 15, 1929, pp. 52-55, 23 figs.

Heat treatment of manganese steel is discussed; results of number of investigations to show effect of treatment at different temperatures; micrographs illustrate variations in grain structure at these temperatures. (To be continued.)

TESTING. The Wear Testing of Manganese Steel, J. H. Hall. *Metal Stampings*, vol. 2, no. 1, Jan. 1929, p. 16.

Development of test for wearing properties of 12-per cent manganese steel to give results comparable with service experience. Brief abstract of paper presented before Am. Soc. Testing Matls.

METALLOGRAPHY

Metallography Simplified for Practical Use in Shop, E. Preuss, G. Berndt and M. v. Schwarz. *Iron Trade Rev.*, vol. 83, no. 26, Dec. 27, 1928, pp. 1624-1626, 10 figs.

Hardened steel is discussed; even with slowest cooling, manganese-hard steel containing about 12 per cent manganese remains in austenitic state and cannot be annealed soft; martensite in iron carbon alloys; sufficient to quench hypereutectoid steels from about 730 to 740 deg. Cent. (Continuation of serial.)

Metallography Simplified for Practical Use in Shop, E. Preuss, G. Berndt, M. v. Schwarz. *Iron Trade Rev.*, vol. 84, no. 2, Jan. 10, 1929, pp. 130-132, 5 figs.

Further discussion of structure of hardened steel; tools showing hardenite after quenching have been hardened excellently; structure similar to hardenite also found in hardened chromium steel and high-speed steel; effects of tempering; strength of steel somewhat reduced by tempering; on account of its great brittleness martensitic structure is not suitable for cutting tools. (Continuation of serial.) Translated from German.

Metallography Simplified for Practical Use in Shop, E. Preuss, G. Berndt and M. v. Schwarz. *Iron Trade Rev.*, vol. 84, no. 4, Jan. 24, 1929, pp. 267-269, 10 figs.

Details and effects of case-hardening steel and proper tempering treatments for case-hardened parts are discussed; examples of effect of quenching on structure of steel. (Continuation of serial.)

Metallography Simplified for Practical Use in Shop, E. Preuss, G. Berndt and M. v. Schwarz. *Iron Trade Rev.*, vol. 84, no. 6, Feb. 7, 1929, pp. 392-394, 7 figs.

Further discussion of case-hardened steel; in order to obtain especially tough cores in case-hardened structural parts, low chromium-nickel steel is used to advantage; good results can be obtained also with ordinary steel if carbon content of part to be case-hardened is not too high; thickness of case-hardened layer for gears; too high temperature or too long a time of carburization can entirely spoil piece. (Continuation of serial.)

METALS

DEFORMATION. Relation of Mechanical Twinning to Plasticity and Strain Hardening (Ueber die Bedeutung der mechanischen Zwillingsbildung fuer Plastizitaet und Verfestigung), T. Schmid. *Zeit. fuer Metallkunde (Berlin)*, vol. 20, no. 12, Dec. 1928, pp. 421-425, 11 figs.

Brief description of both mechanisms of plastic crystal deformation; translation is defined as carrier of large deformations, whereas significance of twinning lies primarily in creation of suitable conditions for further translation; results of tests on zinc and cadmium crystals.

MACHINABILITY. Report on Machinability, E. G. Herbert. *Engineer (Lond.)*, vol. 146, no. 3806, Dec. 21, 1928, pp. 692-695, 18 figs.

Machining processes change physical properties of metals to which they are applied; machinability is not attribute of metals which can be considered or measured apart from particular machining processes; hardness of chips removed by metal-cutting process indicates hardness induced in metal by that process; original hardness of metals gives no indication of machinability; machinability depends on natural angles of built-up edge. Abstract of paper presented before Instn. Mech. Engrs. See also *Engineering (Lond.)*, vol. 126, no. 3284, Dec. 21, 1928, pp. 789-792, 9 figs.

MACHINABILITY. Machinability, E. G. Herbert. *Metallurgist (Supp. to Engineer, Lond.)*, Jan. 25, 1929, pp. 1-2.

Review of paper before Institution of Mechanical Engineers; author studied behavior of series of steels under lathe-cutting conditions, and carried out measurements on hardening effect of cutting operation; it is evident that machinability is much more difficult thing to measure than has hitherto been believed, or even than Herbert's work would,

by itself, still more work he has still more

MACHINABILITY. Test as Indicator of Machinability (Der Bohrvorgang als Arbeitbarkeitskennzahl), S. Patkay. *Symposium on Drilling and Machining of Specimens for Drilling*.

PROPER MECHANICAL PROPERTIES. H. J. W. *Trades Review*, Dec. 28, 1928.

Mechanics of Machining (pendent of innumerable factors each is determined rather than production of metal and metallography).

TEMPERATURE DEVELOPMENT. T. F. D. *Dec. 22, 1929.*

Enumeration of facts in chart is temperature give idea for service of notes structure

MOLTEN METALS. Molten Metals (importance of ten der giorunge vorgang), *Dec. 49-55.*

Measure molten physical molecular their in

METALLURGY. Deter allurgis 28, 1929

Review in Oct. huetter rate change cases, one of short-circuit

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by itself, suggest; considerations raised by paper serve to demonstrate importance of work he has done and need for more work of still more comprehensive kind.

MACHINABILITY — TESTING. Drilling Test as Index of Machinability of Metals (Der Bohrversuch als Kennzeichen der Bearbeitbarkeit der Metalle), G. Schlesinger and S. Patkay. *Werkstattstechnik (Berlin)*, vol. 22, no. 24, Dec. 15, 1929, pp. 677-683.

Symposium of two papers: (1) development of drilling test since 1902; (2) machinability, drilling and twist drills; micro photographs of specimens of steel alloys tested with twist drill.

PROPERTIES. Mechanical Properties of Single Metal Crystals and Crystal Aggregates, H. C. H. Carpenter. *Iron and Coal Trades Rev. (Lond.)*, vol. 117, no. 3174, Dec. 28, 1928, p. 936.

Mechanical properties of metals are dependent on their inner structure; consist of innumerable small crystals; external shape of each is determined by those surrounding it, rather than by its own natural symmetry; production of large crystals; individual crystals and mechanical properties; value of metallography. Abstract of lecture before Wales Inst. of Engrs.

TEMPERATURE EFFECTS. Power Plant Development Limited by Available Materials, T. F. Davey. *Power*, vol. 69, no. 4, Jan. 22, 1929, pp. 129-131, 1 fig.

Enumeration of inherent or physical defects in power-plant construction materials; chart is presented showing effect of temperature on tensile strength of metals; to give idea of applicability of various metals for service throughout plant, résumé is given of notes compiled during design and construction of Long Beach station.

MOLTEN — PROPERTIES. Properties of Molten Metals and Alloys and their Significance for Casting Process (Die Eigenschaften der schmelzflüssigen Metalle und Legierungen und ihre Bedeutung fuer den Gussvorgang), F. Sauerwald. *Gieserei (Duesseldorf)*, vol. 16, no. 3, Jan. 18, 1929, pp. 49-55, 12 figs.

Measurement of physical properties of molten metals and results direct influence of physical properties on casting process; molecular constitution of molten metals and their importance in foundry practice.

METALS TESTING

Determination of Endurance Limits. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 28, 1928, pp. 187-189, 1 fig.

Review of paper by J. Geller published in Oct. 1928, issue of Archiv. fuer das Eisenhuettenwesen, who obtains expressions for rate at which internal energy absorbed changes with frequency for two particular cases, one of tension-compression cycle and one of alternating torsion; other method of short-duration fatigue testing is by measurement of energy absorbed; these methods are criticized by Bohuszewicz and Spaeth in same issue, who propose novel form of which is described.

Symposium in the Kaiser-Wilhelm Institute for Metal Research (Kolloquium im Kaiser-Wilhelm-Institut fuer Metallforschung). *Zeit. fuer die gesamte Giesereipraxis (Berlin)*, vol. 49, no. 51, Dec. 16, 1928, pp. 201-203.

Brief reviews of following papers: Investigation of Brass Crystals, by von Goeler; Aging of Copper and Zinc Alloys, Hansen; Testing of Duralumin, Wassermann; Recrystallization Phenomena of Aluminum, Sachs; Deformation Resistance of Solids, Gels and Liquids, Weissenburg; Growth of Cast Iron O. Bauer.

Metallurgy at the French Aeronautical Exhibition, Galibourg. *Metallurgist (Supp. to Engineer, Lond.)*, Jan. 25, 1929, pp. 8-10, 1 fig.

Brief descriptions of instruments and appliances of definitely metallurgical character included as special feature of Exhibition; testing machines, apparatus for microscopic examination of metals, appliances for use of X-rays, appliances for thermoelectric and magnetic testing, pyrometers, etc., are included. Abstract translated from Revue de Metallurgie, Nov. 1928.

NICKEL ALLOYS

Nickel, Nickel Alloys, and Cobalt. *Metal Industry (Lond.)*, vol. 33, nos. 25 and 26, Dec. 21 and 28, 1928, pp. 583-588 and 607-610.

Dec. 21: Description of works of Henry Wiggin and Co., which begins with pure metals for their melting furnaces and despatch as their final products, mainly sheet, ribbons, and wire; nickel silvers are among most important alloys made by this firm; Ajax-Wyatt frequency induction furnaces used; both hot and cold rolling plants are described. Dec. 28: At Wiggin Street Works laboratories, scleroscope is used for testing of hardness; electrical research laboratories.

NICKEL-CHROMIUM IRON

Nickel-Chromium Alloy Iron, W. R. Shimer. *Am. Mach.*, vol. 70, no. 6, Feb. 7, 1929, pp. 249-250.

Uses and properties of Mayari nickel-chromium iron, especially in automobile manufacture; wear-resisting properties; results of research; no attempt made to change natural proportions of alloy. Abstract of paper presented before Western Metals Congress.

NITRIDATION

Methods of Approximating Certain Physical Characteristics of Nitrided Steel Cases, G. M. Eaton. *Am. Soc. for Steel Treating—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 1-24 and (discussion) 24-35, 18 figs.

Need for inspection of nitrided products is emphasized; vital characteristics of case are hardness and ductility; photomicrographs are given showing impression produced on nitrided case by practically all recognized hardness-testing machines; Vickers diamond indenter hardness tester gives clearest distinction between brittle and ductile characteristics of nitrided cases; warning is given that there is danger of unnecessary failures which will tend to discredit nitrided products in general if heavy production is entered into without adequate inspection testing.

NONFERROUS ALLOYS

HEAT TREATMENT. The Heat Treatment of Nonferrous Alloys, V. O. Homerberg. *Fuels and Furnaces*, vol. 7, no. 1, Jan. 1929, pp. 41-44 and 102, 9 figs.

Wide range of physical properties may be

obtained by heat treatment of nonferrous alloys showing possibility of transformation of alpha plus beta into all beta condition, retention of which can be obtained by quenching from proper temperatures; very finely divided alpha accompanied by marked change in physical properties results from reheating of beta solution.

NONFERROUS METALS

SPECIFICATIONS. Nonferrous Metal Specifications. *Foundry*, vol. 57, nos. 1 and 2, Jan. 1 and 15, 1929, supp. plates nos. 747, 748, 749 and 750.

Jan. 1: Specifications for yellow brass, white-nickel brass, manganese bronze, cast brass to be brazed, and brazing solder are given. Jan. 15: Specifications for hard cast bronze, leaded gun metal, phosphor bronze, phosphor gear bronze, and bronze backing for lined bearings. (To be concluded.)

PYROMETRY

Practical Pyrometry, G. B. Brook, H. J. Simcox and E. Wilson. *Engineering (Lond.)*, vol. 126, no. 3285, Dec. 28, 1928, pp. 822-824, 4 figs.

Subject is divided into three parts; (1) magnetic and electrical insulation; (2) sheathing and protection of couples in furnace; (3) development of quick-reading couple. Paper read before Inst. of Metals.

RAILS

CHROMIUM STEEL. Chromium Steel Rails, T. Swinden and P. H. Johnson. *Rolling Mill J.*, vol. 2, no. 12, Dec. 1928, pp. 521-522 and 534.

Abstract of paper presented before Brit. Iron and Steel Inst., previously indexed from *Engineering*, May 18, 1928.

CORRUGATION. Rail Materials and Corrugation, G. Sproule and C. F. Pascoe. *Elec. Traction*, vol. 24, no. 12, Dec. 1928, pp. 634-635.

Discussion of report on rail corrugation issued by Dept. of Metallurgical Engineering, McGill University; rails are now more resistant and cars lighter and more easily sprung which should cause decline in corrugation troubles; how corrugation starts; use of harder rails; analysis of rail specimens; soundness of steel.

MANGANESE STEEL. Value of Manganese Steel Rails Laid on Curves, K. Kawai and K. Imamura. *Japan Department of Railways—Bul. (Tokyo)*, vol. 16, no. 11, Nov. 1928, pp. 1653-1666.

Life and merits of carbon and manganese-steel rails are considered in connection with amount of wear after 100 days' service; some conclusive remarks are given up on value of latter rails from safety as well as economical viewpoint. (In Japanese.)

MACROSTRUCTURE. The Inverted Macrostructure in Certain Lots of Rails, E. Decherf. *Int. Ry. Congress Assn.—Bul. (Brussels)*, vol. 10, no. 12, Dec. 1928, pp. 1048-1055, 8 figs.

In order to be able to study in logical manner anomalies as macrostructures, following investigations were made: macrostructure tests; arrangement of fibers; investigation into chemical composition; metallographic investigations. From *Revue Universelle des Mines*, Mar. 16, 1928.

METALLURGY. Note on the Inverted Segregation Observed in Certain Rails, A. Portevin. *Int. Ry. Congress Assn.—Bul. (Brussels)*, vol. 10, no. 12, Dec. 1928, pp. 1073-1075, 3 figs.

Brief recapitulation of what "inverted segregation" consists; mechanism of phenomenon. From *Revue Universelle des Mines*, June 1928.

STEEL—MANUFACTURE. Heat-Treated Rails Make Slow Progress, H. B. Greensted. *Can. Machy. (Toronto)*, vol. 39, no. 24, Nov. 29, 1928, p. 35.

Rail-steel manufacture and specifications discussed; development of rail manufacture; precautionary nick and break test employed by Canadian Railways; prospect of using heat-treated rails. Brief abstract of paper presented before Montreal Chapter of Am. Soc. Steel Treating.

TRANSVERSE FISSURES. Surface Transverse Fissures in Rails, C. Dantin. *Int. Ry. Congress Assn.—Bul. (Brussels)*, vol. 10, no. 12, Dec. 1928, pp. 1041-1047, 5 figs.

Description of defects; method of formation; reproduction of fissures by deliberately slipping wheels; investigation into fissured rails; drop and bend test; non-treated and heat-treated rails. Abstract translated from *Génie Civil*, Jan. 21, 1928.

REFRACTORY MATERIALS

Progress in Refractory Materials, A. B. Searle. *Metal Industry (Lond.)*, vol. 34, no. 3, Jan. 18, 1929, pp. 71-74.

Progress is being made by manufacturers in improved selection of materials, more effective treatment prior to shaping goods, and more skilled control of operation of drying and burning; use of tunnel kiln in which goods are burned on cars, has led to much greater regularity in burning; zinc smelting; principal cause of disintegration in zinc retorts; Continental conditions; lead and copper working; brass and bronze melting; aluminum and antimony furnaces; furnaces for precious metals; nickel and muffle furnaces.

SPECIFICATIONS. A Manufacturer's Viewpoint Regarding Specifications, G. A. Balz. *Am. Refractories Inst.—Tech. Bul.* no. 26, Aug. 1928, 4 pp.

Specifications are most essential fact in progressive industrial development; water-spalling test in the writer's opinion should be discarded; specifications should be carefully prepared after thorough deliberation; and should be amenable to revision from time to time as practical experience may indicate.

SPRING STEEL

HEAT TREATMENT. On the Heat-Treatment of Spring Steel of Globular Cementite Structure, S. Kokado. *Soc. Mech. Engrs.—Jl. (Tokyo)*, vol. 31, no. 138, Oct. 1928, pp. 954-967, 5 figs.

In this paper, influence of heating time on tensile strength of annealed as well as of quenched and tempered fine spring steels (0.45 mm. thick) are studied. (In Japanese.)

SPRING TESTING MACHINES

A High-Speed Endurance Testing Machine for Leaf Springs, J. Bradley. *Engineering (Lond.)*, vol. 127, no. 3287, Jan. 11, 1929, pp. 36-37, 3 figs.

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ratory for Springs Research Committee; in designing machine, following features were regarded as essential: (1) speed should be not less than 1000 cycles per minute; (2) test piece should be of form requiring little or no preliminary machining; (3) it should be subjected to uniform bending moment over length of several inches; (4) stress applied should be independent of speed of machine.

STAINLESS IRON

TEMPERATURE EFFECT. Effect of Temperature on Stainless Iron, R. Sergeson. *Heat Treating and Forging*, vol. 15, no. 1, Jan. 1929, pp. 55-59, 11 figs.

Study of physical properties of high nickel-chromium alloy of iron under various thermal treatments; proper conditions for utility are indicated; izod impact tests; effect of cold work and of normalizing temperature on cold-drawn bars.

STAINLESS STEEL

Some Applications of Stainless Steel. *Metallurgist (Suppl. to Engineer, Lond.)*, Dec. 28, 1928, pp. 190-191.

Cutlery is most natural application of stainless steel; there are two important further applications in mechanical construction which deserve special mention, namely, pistons for pumps or hydraulic presses and turbine blades; these are discussed.

STEEL

ANTI-CORROSIVE. Rust-, Acid-, and Heat-Resisting Steels. *Iron and Steel of Canada (Gardenvale, Que.)*, vol. 11, no. 12, Dec. 1928, pp. 377-379.

Six classes of steels shown in table are ones now most widely used; heat-resisting steels; theory of corrosion.

BRONZE EFFECT. Effect of Molten Bronze on Steel, I. Hook. *Acetylene J.*, vol. 30, no. 7, Jan. 1929, pp. 277-280, 9 figs.

Paper discusses effect of flowing molten bronze upon steel; liabilities incurred; and conditions governing effect of one upon other; importance of stress on contact surface. Paper read before Int. Acetylene Assn.

CARBON SOLUBILITY. Solubility of Carbon in Normal and Abnormal Steels, O. E. Harder and W. S. Johnson. *Am. Soc. for Steel Treating—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 49-62 and (discussion) 62-68, 34 figs.

Continuation of study on normal and abnormal steels in Vol. 13, 691-1008 of *Trans.* present study of solubility of carbon in typical normal and abnormal steels, making use of longer carburizing times and different carburizing temperatures; results show that for given carbon content precipitation takes place in abnormal steel at higher temperature; as carburizing temperature increases amount of carbon taken up by normal and abnormal steels becomes essentially same.

HARDENING. Recent Investigations of the Theory of Steel Hardening (Neuere Untersuchungen ueber die Theorie der Stahlhaertung), E. Scheil. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 6, Dec. 1928, pp. 375-387 and (discussion) 387-388, 39 figs.

Conversion of austenite into martensite; influence of pressure and stresses; change of austenite with annealing; determination of austenite content; Maurer's hardening hypothesis; influence of grain size on hard-

ness; hypothesis of Smekal and Hanemann; austenite-martensite diagram of state.

HARDENING. The Constitution of Steel and Cast Iron, F. T. Sisco. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 158-169, 1 fig.

Present installment takes up in introductory manner hardening of steel; operations of heating for hardening and holding at hardening temperature are discussed, followed by description of effect of speed of cooling upon location of lower or transformation point, which, from standpoint of stable equilibrium, marks phase change from austenite to pearlite. (Continuation of serial.)

HEAT TREATMENT. Nitriding Is Hailed as Important Advance in Heat Treating. *Iron Trade Rev.*, vol. 84, no. 1, Jan. 3, 1929, pp. 28-29, 1 fig.

Progress in development of Aluminum-chromium steel and in art of nitriding them has been most noticeable during past year; opinions of leading authorities quoted; economic aspects of nitriding situation; development of furnaces, both electric and gas fired; progress made along lines of theory and research; improvement of furnace and temperature control.

HEAT TREATMENT—TEMPERING. The Tempering of Steel. *Chem. Age (Lond.)*, vol. 20, no. 497, Jan. 5, 1929, pp. 2-3, 2 figs.

Discussion of Homo method; transmission of heat; methods of heating.

METALLOGRAPHY. Microscopic Studies of an Inactive Stratum in Carbon Steel and Etching Structure Caused Thereby (Mikroskopiska studier av ett passiverande skikt a kolstal och darav fororsakad etsstruktur), C. Benedicks and P. Sederholm. *Jernkontorets Annaler (Stockholm)*, vol. 83, no. 11, Nov. 1928, pp. 565-573 and (discussion) 573-575, 11 figs.

Statement by Metallographic Institute on original experiments on etching of particularly homogeneous carbon steel with highly diluted nitric acid.

TEMPERATURE EFFECT. The Effect of Furnace Atmospheres on Steel, R. G. Guthrie. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 1, Jan. 1929, pp. 96-111 and (discussion) 111-116, 11 figs.

Tests with atmospheres commonly met within heat-treating furnaces were conducted at temperature of 1500 deg. Fahr. as being an average temperature for heat treatment of carbon steels of eutectoid composition; author uses series of photomicrographs in which type of steel, conditions of time, temperature, pressure, velocity, etc., are constant with only variable being different simple gases; an additional set of photomicrographs show effects of mixing gases in various proportions; chemistry of combustion of gas is added as appendix.

TEMPERATURE EFFECT. Strength of Low Carbon Steels at High Temperatures, R. G. Batson. *Engineer (Lond.)*, vol. 146, no. 3807, Dec. 28, 1928, pp. 707-709, 8 figs.

Work described was carried out at Engineering Department of National Physical Laboratory, undertaken at request of Engineer Surveyor-in-Chief to Board of Trade; results of tests on steel up to approximately 650 deg. Cent.; results of tests on 0.10 per cent carbon steel between 500 and 650 deg. Cent. are also given; general discussion of "creep" results.

TENSILE TESTING. Strain Markings in Mild Steel Under Tension, H. S. Rawdon. *Metal Stampings*, vol. 1, no. 7, Dec. 1928, pp. 525-526.

Abstract of report in U. S. Bur. of Standards—Research paper, no. 15, previously indexed from Sept. 1928, issue of their JI. of Research.

X-RAY ANALYSIS. Nature of "Blue Brittleness" in Steel, K. Honda. *Iron Age*, vol. 122, no. 26, Dec. 27, 1928, p. 1638.

Brief abstract of paper previously indexed from Tohoku Imperial Univ.—Sci. Reports (Sendai), July 1928.

YIELD POINT. The Physical Significance of the Yield Point. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 28, 1928, pp. 186-187.

Review of two articles recently published in German technical press; first by W. Kuntze and G. Sachs in *V. D. I. Zeit.*; July 21, 1928, describes experiments carried out with new silicon steel, which has well-defined yield point; authors interpret their results as affording strong indications of presence of brittleness; second paper, by Kuehnelt, *V. D. I. Zeit.*, Sept. 1, 1928, deals with importance of yield point to designer.

STEEL ANALYSIS

Method of Routine Analyses in Metallurgical Works (Les méthodes d'analyses courantes dans les usines métallurgiques), L. Persoz. *Revue de Chimie Industrielle (Paris)*, vol. 37, no. 444, Dec. 1928, pp. 382-384, 2 figs.

Determination of oxygen in steel; description of apparatus and its operation.

TROOSTITE DETERMINATION. Troostite, A. E. Cameron and I. E. Morrison. *Roy. Soc. Canada—Trans., Section 3*, 1928, pp. 289-291, 17 figs. on supp. plates.

Short paper on microscopic examination of specimens of medium carbon (0.5 per cent C) and high carbon (0.8 per cent C) steels after series of heat treatments and quenching operations, followed by tempering and drawing operations at temperatures of 100, 200, 300, 400, and 500 deg. Cent.

STEEL CASTINGS

Hot Cracks in Steel Castings (Warmrisse im Stahlformguss), K. Singer. *Giesserei (Duesseldorf)*, vol. 15, no. 49, Dec. 7, 1928, pp. 1225-1229, 9 figs.

Aim of present work is to find means and methods of achieving greater safety in production of difficult steel castings which have tendency to crack.

Symposium on Steel Castings. *Foundry Trade JI. (Lond.)*, vol. 40, nos. 648 and 649, Jan. 17 and 24, 1929, pp. 53-54 and 56, and 72-74, 1 fig.

Jan. 17: Present and future position of steel castings industry; steel castings for merchant marine; futility of hammering test; Admiralty specification deemed to be low; light-weight valves and heavy brackets exhibited; homogeneity of steel castings; steel gearing for sugar machinery. Jan. 24: Elastic and plastic strains; strength of elevated temperatures; properties of electric steel; oil-sand cores and hot cracks.

METALLURGY. Metallurgy of Steel Castings, A. McCance. *Foundry Trade JI. (Lond.)*, vol. 39 and 40, nos. 645 and 647, Dec. 27, 1928, and Jan. 10, 1929, pp. 465-466, and 25-26, 15 figs.

Main difficulties in manufacture of steel castings are concerned with problems of contraction; liquid contraction or contraction of molten metal in mold down to freezing temperature; freezing contraction which takes place when metal passes from liquid to solid condition; and cooling contraction which takes place during cooling down of casting from solid condition at high temperature to condition at ordinary temperatures. Paper presented before Inst. Brit. Foundrymen and Instn. Mech. Engrs.

STEEL CEMENTATION

Metallic Cementation, Laissus. *Metallurgist (Supp. to Engineer, Lond.)*, Jan. 25, 1929, pp. 7-8.

Cementation with uranium is taken up by author who also summarizes his conclusions arrived at in previous seven papers which deal with zirconium, titanium, vanadium, chromium, boron, molybdenum, tungsten, cobalt, and tantalum; author bases much of his interpretation of microstructures which he finds in "cemented" layers upon equilibrium diagrams of alloy systems in question. Abstract translated from *Revue de Métallurgie*, Jan. 1928.

STEEL FOUNDRY PRACTICE

Two Developments in Steel Casting, J. H. Hall. *Iron Age*, vol. 123, no. 1, Jan. 3, 1929, pp. 45-46.

Better product to meet severe demands of industry; cooperative advertising success; larger use of alloy castings; autogenous welding as competitor; growing demand for construction machinery.

STEEL HEAT TREATMENT

Design in Relation to Heat Treatment, R. T. Bayless. *Iron Age*, vol. 123, no. 5, Jan. 31, 1929, pp. 341-342.

Necessity for closest kind of cooperation between designer of heat-treated steel parts and heat treater; what causes distortion and cracking; fundamental principles of good design from heat-treatment standpoint; any design which causes concentration of load at certain point or along sharp line may cause start of failure; hardening cracks; sharp internal angles should be avoided. Use of generous fillets recommended. Abstract of paper presented before Western Metal Congress.

STEEL INGOTS

The Production of Certain Steel Ingots, E. Lubojatzky. *Rolling Mill JI.*, vol. 2, no. 12, Dec. 1928, pp. 531-533.

Relation between cooling curves and structures of certain alloy-steel; solidification of steel ingots. (Continuation of serial.) Abstract translated from *Montanistische Rundschau*.

MANUFACTURE. The Manufacture of Acid Open-Hearth Steel for Forging Ingots, H. P. Rassbach. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 2, Feb. 1929, pp. 289-302, 2 figs.

Recent demand for heavy forgings requiring high physical properties; reasons for selection of acid open-hearth steel for heavy forgings; suitable charges for acid furnace; operation of heat is discussed and reasons are given for various procedures; special emphasis is placed on desired condition of

bath and slag in different stages of heat; composition, size of ingot, and use for which it is intended are factors that determine method of pouring to be employed.

STEEL MANUFACTURE

RIMMED STEEL. Manufacture of Thick-skinned Ingots of Rimming Steel, H. D. Hibbard. *Fuels and Furnaces*, vol. 7, no. 1, Jan. 1929, pp. 67-69.

Influence of melting and pouring temperatures on quality of ingots of rimming steel and procedure for obtaining proper temperatures, finish of heat, and methods of castings into molds. (Continuation of serial.)

STEEL METALLOGRAPHY

MARTENSITE STRUCTURE. Further Observations on the Microstructure of Martensite, F. F. Lucas. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 2, Feb. 1929, pp. 339-365 and (discussion) 365-367, 15 figs.

Description of number of quenching and tempering experiments in which commercial high-quality tool steels were used; representative structures found in the quenched and various tempered conditions are illustrated and discussed. Bibliography.

PHOTOGRAPHY. Photographs on Solid Steel, L. V. Dodds. *Metallurgist (Supp. to Engineer, Lond.)*, Jan. 25, 1929, p. 16.

Method whereby image is projected into solid block of hardest chrome steel under agency of high-explosive nitro starch; known as "Munroe Effect," being discovery of C. E. Munroe, veteran explosives expert of U. S. Bureau of Mines; record is obtained which is as permanent as steel itself; examples of this method of photography; commercial applications are being found speedily.

RECRYSTALLIZATION. Recrystallization Occurring with Hot Rolling (Die Rekristallisation beim Warmwalzen), W. Tafel, H. Hanemann and A. Schneider. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 1, Jan. 3, 1929, pp. 7-12, 13 figs.

Tests on recrystallization of soft steel after rolling under different compressors at 650 to 1200 degrees Cent.; preliminary tests to determine best spots for measurement of grain size, influence of initial grain size, and difference between grain size in transverse and longitudinal cross-section; three-dimensional diagram is plotted, results are discussed, and conclusions drawn.

STEEL QUENCHING

The Quenching of Steel, P. J. Haler. *Engineering (Lond.)*, vol. 127, no. 3286, Jan. 4, 1929, pp. 25-26, 8 figs.

View is advanced that warping is largely due to action of quenching fluid, which produces differential cooling on various faces; it has been shown that, in severe quenching, periphery of cube is always softer than central portions, this anomalous phenomenon being explained by presence of arrested austenite in martensite; it is suggested that uneven cooling, as explained in this paper, causes this phenomenon.

SURFACE COOLING. Surface Cooling of Steels in Quenching, H. J. French, G. S. Cook and T. E. Hamill. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 2, Feb. 1929, pp. 217-284 and (discussion) 284-288.

Study of surface cooling of steel spheres when quenched in water, sodium-hydroxide solutions, oils or cooled in air; effects of some of variables encountered in commercial heat treatment upon cooling of steel bodies; center and surface-cooling curves obtained for pressure spray quenching with water are correlated with tensile and impact properties obtained in low carbon steels, American ingot iron, and wrought iron.

STEEL TESTING

Ehn Carburizing Test (Ein Beitrag zur Kenntnis der Ehschen Zementationsprobe), R. Wasmuth and P. Oberhoffer. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 3, Jan. 17, 1929, pp. 74-77, 15 figs.

Result of attempt to apply carburizing test to investigating and distinguishing between good and bad steels while plant is in operation; influence of deoxidation on specimen; phenomena occurring with decarbonization.

STRIP MILLS

COLD ROLLED. Adds Cold-Rolled Strip Department. *Iron Age*, vol. 122, no. 26, Dec. 27, 1928, pp. 1629-1631, 4 figs.

Description of new pickeling and galvanizing departments and cold-rolled strip plant of Sharon Steel Hoop Co.; skillful tie-up with older sections; remarkable ease with which steel coming from hot mills is transferred; two sets of continuous pickling tanks; strip pulled through tanks at constant speed by automatic control; three sets of tandem roughing mills with reels and coilers; all mills equipped with roller bearings.

TUBES, STEEL, SEAMLESS

MANUFACTURE. Contrasts Operation of Pilger and Plug Type Tube Mills, E. Roeber. *Iron Trade Rev.*, vol. 84, no. 2, Jan. 10, 1929, pp. 133-134.

Discussion of paper, "The Manufacture of Seamless Tubes", presented by R. C. Stiefel and G. A. Pugh before American Society Mechanical Engineers; maximum length and wall thickness are factors in considering advantages of two processes from standpoint of output and costs; outputs not comparable. From *Stahl und Eisen*, Sept. 6, 1928.

The Manufacture of Seamless Steel Tubes, H. E. Passmore. *Ry. Club Pittsburgh—Official Proc.*, vol. 28, no. 1, Nov. 22, 1928, pp. 5-25 and (discussion) 25-28, 7 figs.

Mannesmann machine for piercing round billets embodies principle of diagonal rolling; operation of piercing; principles involved in forming cavity; flow of metal in piercing; function of mandrel; Steifel machine; operation and action of mill; comparison of processes; general description of present-day practice; some of uses and materials used in making seamless tubes; variations or tolerances that can be maintained.

TUNGSTEN ALLOYS

Tungsten-Carbon System (Ueber das System Wolfram-Kohlenstoff), K. Becker. *Zeit. fuer Metallkunde (Berlin)*, vol. 20, no. 12, Dec. 1928, pp. 437-441, 9 figs.

Metallographic and X-ray investigation gives proof of carbide phases in molten tungsten carbon and in tungsten wires carburized from gas phase; electric conductivity of carburized tungsten wires.

Recommended Practice Committee Releases

CONSTITUTION OF NICKEL-ALUMINUM ALLOYS

N. B. Pilling* and T. E. Kihlgren*

Our sole information on the constitution of the nickel-rich alloys with aluminum comes from Gwyer¹, whose diagram, reproduced in part as shown

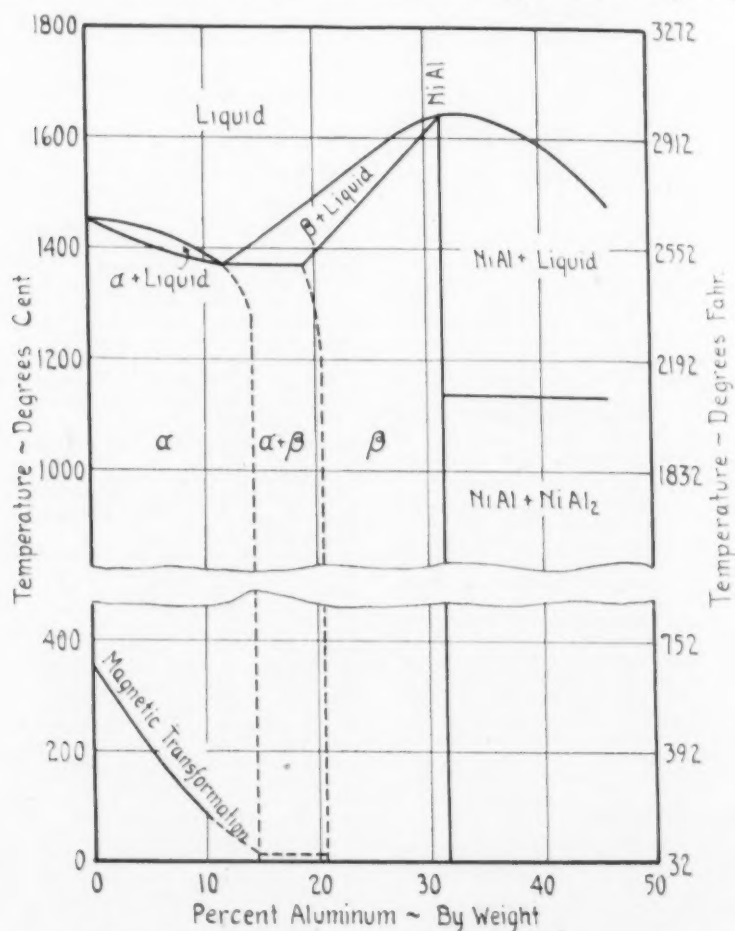


Fig. 1—Nickel-Aluminum Equilibrium Diagram (Gwyer).

in Fig. 1, was published in 1908. Nickel and aluminum are completely miscible when fused, although a compound, NiAl, is formed which has a melting point nearly 200 degrees Cent. greater than that of nickel. About 14

*The International Nickel Co., Inc., Bayonne, N. J.

Prepared for the Nonferrous Data Sheet Committee of the Institute of Metals Division of the A. I. M. E. and the Recommended Practice Committee of the A. S. S. T.

per cent aluminum is taken up in solid solution in nickel, forming the Alpha phase, while the compound NiAl may form solid solutions with as much as 10 per cent nickel, although not with aluminum.

The indicated changes in solid solubility with temperature of the α and β phases were based on a few microscopic examinations of specimens quenched from the vicinity of the solidus temperature and may need to be taken with some reserve.

The magnetic transformation of nickel is depressed by aluminum. The β phase is non-magnetic.

Reference

1. A. G. C. Gwyer; "Zeitschrift für Anorganische Chemie," Vol. 57, page 113, 1908.

CASE DEPTH MEASUREMENT*

This committee was organized to analyze and study the various methods of indicating the depth of case of carburized steels and if found practical to recommend a standard method of case depth measurement.

It must first be recognized that when steel is carburized the "total case depth" comprises the steel in which the carbon content is higher than in the original material or higher than the carbon content of the core. When the carburized steel is subsequently heat treated, a certain percentage of the case is high enough in carbon to permit hardening, thus fulfilling the purpose of carburizing. This useful portion of the case we shall designate as the "effective case." The properties of this effective case are necessarily quite different for parts subjected to various uses. Thus, in a part subject only to wear or abrasion, the effective case is relatively small and consists only of that portion above a hardness which will resist the wear encountered. On the other hand, steel case hardened to increase strength, and which must also possess toughness, will have a deeper effective case and will comprise all the case above a much lower hardness value.

Methods for Measuring Case Depth—There are many methods in everyday use for measuring case depth. Most of these methods indicate a depth of case between the effective and total depths. They are dependent upon the state of heat treatment at the time of measurement, the carbon gradient in the case, and many other factors which will be discussed later. The several methods for measuring case depth are as follows:

1. Scale measurement of ordinary fracture.

Note—See note under method 3.

2. Scale measurement of blued fracture.

The bluing is obtained by slowly heating the fractured specimen until the temper colors appear. The higher carbon sections of the case reach a

*A tentative report of the Sub-Committee of the Recommended Practice Committee of the A. S. S. T. on Measuring Case Depth. The membership of the Sub-Committee was as follows: Stanley P. Rockwell, Chairman; B. F. Shephard, Dr. R. W. Woodward, and R. J. Allen.

Criticisms of this report are solicited and should be directed to J. E. Donnellan, Secretary of the Committee, 7016 Euclid Avenue, Cleveland, Ohio.

blue color before the lower carbon sections. Bluing is best done on a hot plate.

Note—See note under method 3.

3. Scale measurement of ordinary fracture which has been dipped in acid.

Nitric acid is usually used, either strong or diluted with water or alcohol. The high carbon section of the case is colored before the low carbon section.

Note—Methods 1, 2, and 3, which measure depth by reference to fracture, are subject to considerable variation, depending upon the heat treatment and subsequent carbon diffusion, and by the personal equation of the operator conducting the test. These methods closely approximate the total depth of case. The usual crystalline cores are more difficult to estimate than those having amorphous or fibrous cores. Alloy steels show less demarcation between case and core than do carbon steels. Of the three methods, method number 2, fracture bluing, seems to offer the clearest demarcation between case and core, and thus offers less error in measuring the total case depth.

4. Scale measurement of polished cross sections.

(a) Blued as in method 2.

(b) Etched as in method 3.

Note—This method is subject to similar errors as 1, 2, and 3, but eliminates some of the errors possible in the latter methods.

5. Microscopic measurement of highly polished cross sections. Etched in standard reagents.

Note—The depth may vary with this method, due to the observer's ideas as regards what percentage of carbon makes up the case, and how much increase in carbon is allowable for the core. The division between the eutectic zone and the hypoeutectic is readily discerned and probably bears a close relation to the "effective case" depth. This method requires special equipment and trained personnel.

6. File or drill through from the core at right angles to the case until the hardness of the case resists further cutting or abrasion.

Measurement of the remaining hard case is taken as the depth. This method depends on the removal, say by grinding, of enough of the case so that a drill or file may be used.

Note—This method depends upon the use of a drill or file and is subject to the personal equation.

7. Machine off successive amounts of surface from soft carburized work. Analyze and base results of depth on individual conception of the amount of carbon which can be considered case.

Note—This method of chemical analysis is quite costly and is subject to the same limitations as method 5.

8. Weigh the parts to be carburized before and after carburizing, and compute the added carbon and depth of penetration.

Note—This method has the advantage of being non-destructive but involves complicated calculations.

9. Magnetic Analysis.

Note—Magnetic methods are non-destructive and are promising, but it is felt that more research is necessary before these methods are available for commercial routine.

10. Taper-grind hardened work so that some unit of lineal measurement of surface represents a known depth from original surface, and test each unit of lineal surface by some practical form of hardness test.

This not only shows depth, but characteristics of the different depths from the surface in terms which are readily convertible to resistance to wear and deformation.

Note.—Taper-grinding offers the simplest way of determining effective case depth. It further shows the characteristics of the hardness of the total case at different depths and checks the carbon distribution in a fairly close way.

Results and Recommendations—Table I shows the results of a few different methods of measurements on several typical samples. These samples are the same ones used in an investigation by Rockwell and Downes which

Table I
Depth Measurements of Different Methods

Steels S. A. E. Nos.	Heat Treated Condition	Plain Fracture Scaled Method 1 Inches	Etched Fracture Scaled Method 3 Inches	Etched Cross-Section Microscope Method 5 Inches	Taper Grind, Hardness over C60 Rockwell Method 10 Inches
1015	Pot quenched032	.031000
	Pot cooled029	.029
	H. T. for case034	.032	.032	.020
	H. T. for case and core.	.036	.031	.032	.022
2315	Pot quenched036	.037	.036	.018
	Pot cooled031	.032	.030
	H. T. for case036	.037	.036	.018
	H. T. for case and core.	.045	.042	.041	.010
3115	Pot quenched036	.035	.035	.018
	Pot cooled030	.030	.030
	H. T. for case043	.039	.040	.022
	H. T. for case and core.	.046	.039	.040	.016
2512	Pot quenched028	.030	.032	.018
	Pot cooled016	.019	.031
	H. T. for case038	.034	.034	.016
	H. T. for case and core.	.038	.034	.033	.022
6120	Pot quenched028	.033	.033
	Pot cooled017	.020	.020
	H. T. for case031	.033	.033
	H. T. for case and core.033	.034

were reported in the TRANSACTIONS of the A. S. S. T., Volume V, page 285, and which gave depth measurements by method 1, 3, and 5. Measurements by method 10 have subsequently been made by a number of the sub-committees in which the samples were taper-ground and Rockwell "C" hardness readings taken along a taper of 0.030 of an inch per inch.

These data show remarkably well the influence of the heat treatment condition on case depth measurements and also the effect of various types of steels. Methods 1, 2, and 3 give results which for all practical purposes can be considered the total case depth when performed on samples quenched directly from the carburizing heat. Method 2 on measuring the blued fracture gives consistent results and can be readily performed in the heat treating department. This method is recommended for obtaining the total depth of case.

Method 10 consisting of hardness measurements on a tapered section is recommended for obtaining the effective case depth. As the different hardness testing devices exert more or less loads on the work which could cause the case to break down into the core, mention must be made of the testing machine used for this test.

The sub-committee therefore recommends that the measurement of case depth shall consist of two figures, one giving the total case as measured by a scale on a blued fracture; the second, the effective case or depth having a hardness figure above a prescribed minimum. Such results would be as follows:

Depth of case:

Total—0.025 of an inch.

Effective—0.010 of an inch over Rockwell C58.

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